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STUDY OF CRACKING PERFORMANCE IN SMALL SCALE RESTRAINT TESTS

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*Final Report
by*

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EXECUTIVE SUMMARY

Cracking tests, such as the Welding Institute of Canada small scale restraint test, are widely used to evaluate the sensitivity of steels to hydrogen assisted cracking. Cracking in such tests, however, often occurs in the weld metal rather than the heat affected zone, which leads to ambiguous results when the test is used to rank steels. The present program was carried out to determine if certain test parameters have an effect on the location of cracking with a view to specifying test procedures that would ensure cracking occurred in the region (weld metal or heat affected zone) that was being evaluated.

A simple one-dimensional analysis of the stress and hydrogen distribution as a function of time was conducted and showed that lowering the restraint (by increasing the restraint length of the test) would delay the onset of reaction stress. The analysis also showed that the peak hydrogen level in the heat affected zone was delayed. Thus lower restraint levels should favour heat affected zone cracking and high restraint levels, weld metal cracking.

Numerical modelling using the finite element method (FEM) did not show any significant effect of restraint on delay of stress but did confirm the delay in peak hydrogen. The FEM analysis did not suggest an effect of restraint or preheat on cracking location. A limited number of experimental tests was carried out which did not show any significant effect of test parameters on crack location.

SOMMAIRE EXECUTIF

Les essais de fissuration, tels que les essais de sollicitation longitudinale à petite échelle effectués à l'Institut de soudage du Canada, sont largement utilisés pour évaluer la sensibilité des aciers à la fissuration sous hydrogène. Toutefois, dans ces essais, la fissuration se produit souvent dans le métal de soudure plutôt que dans la zone affectée par la chaleur, ce qui conduit à des résultats ambigus lorsqu'on utilise l'essai pour classer des aciers. Le présent programme a été réalisé dans le but de déterminer si certains paramètres de l'essai ont un effet sur l'emplacement de la fissuration avec pour objectif de définir des méthodes d'essai qui permettraient d'obtenir la fissuration dans la zone (métal de soudure ou zone affectée par la chaleur) faisant l'objet de l'évaluation.

Nous avons effectué une analyse unidimensionnelle simple de la distribution de la contrainte et de l'hydrogène en fonction du temps et nous avons montré qu'en réduisant la sollicitation longitudinale (en augmentant la longueur de sollicitation de l'essai), on retardait l'apparition de la contrainte. L'analyse a aussi montré que le niveau d'hydrogène dans la zone affectée par la chaleur mettait plus de temps à atteindre sa valeur maximale. Par conséquent, une sollicitation longitudinale plus faible favorise la fissuration dans la zone affectée par la chaleur et une sollicitation longitudinale plus élevée favorise la fissuration dans le métal de soudure.

Une modélisation numérique basée sur la méthode des éléments finis (MEF) n'a révélé aucun effet notable de la sollicitation longitudinale sur le retard de la contrainte, mais elle a confirmé le retard dans l'apparition de la pointe d'hydrogène. D'après les résultats de l'analyse MEF, la sollicitation longitudinale ou le préchauffage n'ont pas d'effet sur l'emplacement de la fissuration. Nous avons réalisé un nombre limité d'essais expérimentaux qui n'ont révélé aucun effet important des paramètres d'essai sur l'emplacement de la fissuration.

STUDY OF SMALL SCALE RESTRAINT CRACKING TEST

Introduction

One of the most serious defects in welds is hydrogen assisted cold cracking, and considerable effort has been expended in determining the factors that influence cracking and in developing tests for evaluating the susceptibility of materials. In the past, with relatively high carbon steels, cracking invariably took place in the heat affected zone which had a greater propensity to form martensite and was usually much harder than the weld metal. With the development of modern low carbon steels, the susceptibility of the heat affected zone has been considerably lowered, and now when cracking occurs it is just as likely to occur in the weld metal as the heat affected zone. In many practical cases it is difficult to identify the exact point of initiation, and cracks may appear in the heat affected zone or the weld metal, or both, depending on the section being observed.

This has created a problem in using small scale tests such as the Welding Institute of Canada (WIC) small scale restraint test to rank the susceptibility of steels to hydrogen cracking. If cracking occurs predominantly in the weld metal in the test, then steels would be ranked according to their weld metal susceptibility which could be misleading unless the test exactly reproduced the form of cracking experienced in practice. Furthermore, the location of cracking in a small scale test does not simply depend upon the inherent susceptibility of the heat affected zone and the weld metal but may depend on the procedural parameters of the test. There is, therefore, a need to study the ways in which procedural parameters affect the location of cracking in small scale tests to ensure that tests faithfully reproduce conditions in practice when test results are to be taken at face value and also to modify test parameters to induce cracking solely in the heat affected zone or weld metal when a general evaluation of materials is required.

In the present program the WIC small scale restraint test has been studied since this has proven useful for evaluating crack susceptibility in modern, low carbon pipeline steels and has been increasingly used by major companies in prepurchase evaluations. The aim of the program was to determine whether test parameters might

affect the stress and hydrogen distribution in such a way that it could control the location of cracking. Effects of inherent susceptibility due to composition and microstructure were not investigated in this program.

Literature Survey

A literature survey was carried out using standard databases supplemented by a review of International Institute of Welding (IIW) documents. Most of the relevant papers came from the latter source. The majority of work has been directed at heat affected zone cracking, and test designs have often been specifically chosen to force cracking in the heat affected zone. For example, the extensive work of Ito and Bessyo at Sumitomo Steel [1] in the late 60's used the Tekken test with a sloping Y preparation. In their work, cracking always initiated in the heat affected zone [2] although cracks could run up into the weld metal. However, weld metal cracking is more likely to occur in a Tekken test when a straight Y preparation is used. In the fillet weld controlled severity test (CTS test), widely used in the U.K. and the basis of the British guidelines for safe welding procedures [3], cracking may occur in both the weld metal and the heat affected zone, but the extensive testing at the Welding Institute [4] revealed that when preheat increased, weld metal cracking was eliminated before heat affected zone cracking. Thus for the steels examined in that program, HAZ cracking determined the safe welding conditions in the CTS test.

Graville [5] classified the types of cracks observed in CTS tests on a quenched and tempered steel. Both the gap dimensions and the preheat influenced the location of cracks and the results are summarized in Fig. 1. In this work it was found that the depth of gap affected the location of cracking and the effect was attributed to poor heat transfer across the gap. Although not recognized at that time, the possibility exists of the poor heat transfer causing greater amounts of contraction at lower temperatures which could also influence the position of cracks as will be discussed later. Critical preheats at which cracking was eliminated were controlled by root cracking.

In the small scale restraint test developed at the Welding Institute of Canada [6], the joint preparation may be selected to represent that being used in practice. In early work [7] a single vee with the weld metal at the mid-thickness location was used, and

cracking usually initiated in the heat affected zone and progressed into the weld metal. The subsequent use of preparations representing pipe welds in which the root run at the bottom of the groove created eccentricity resulted in predominance of cracking in the weld metal. This latter configuration is believed to represent conditions in practice when lifting strains are applied [8].

Sawhill [9] used the WIC test with both single bevel preparations and double bevel preparations. His experience indicated that the single bevel preparation generally encouraged cracking to initiate in the heat affected zone.

Until the last few years, there does not seem to be any comprehensive study on the criteria for location of cracking. It has generally been assumed that heat affected zone cracking dominates, which is probably true for the higher carbon steels welded with low and medium strength electrodes. In one early study, Interrante et al. [10], suggested that weld metal cracking might occur when its hardness exceeded that of the heat affected zone. However, studies [11] of weld metal cracking using tests such as the G-BOP test, indicate that weld metal hardness is an unreliable indicator of crack susceptibility.

In developing new models for predicting necessary preheats for avoiding cracking, Suzuki and Yurioka at Nippon Steel [12] explored the idea of local accumulation of hydrogen. Their analysis showed that when the strain concentration in the heat affected zone is taken into account, higher levels of hydrogen may result there than would be expected from standard diffusion theory (Fick's Law). They were able to take this into account in a comprehensive model by relating the strain to the degree of external restraint on the joint. Their model, however, does not indicate where the cracking occurs since again it is presumed that the heat affected zone dominates. Nevertheless, the local distribution of hydrogen could be an important factor in determining where cracking occurs, and in the present work the macroscopic distribution of hydrogen is considered.

The only work that specifically addresses the question of the location of cracking is that of Matsuda et al. [13, 14]. Matsuda considered the time after welding at which the stress and hydrogen concentration would reach a critical value. They used the

tensile restraint cracking (TRC) test in which the load can be applied at specific times after welding. Careful analysis of the fracture face allowed them to classify cracking in four modes depending on the degree to which initiation in the heat affected zone contributed to the total crack. They found that with a high applied stress and high diffusible hydrogen content or high hardness, crack initiation was in the weld metal. If loading in the TRC test for HY 130 steel was at 150°C, then cracking initiated in the weld metal. If loading was delayed and applied after the weld cooled to room temperature, the location of crack initiation moved to the heat affected zone. They were able to explain the results by considering the hydrogen and stress in the weld metal and the heat affected zone as functions of time (Fig. 2). The hydrogen concentration in the weld metal is initially high and decays steadily with time after welding. In the heat affected zone, however, the hydrogen concentration is zero, but it rises to a peak as hydrogen diffuses across the fusion boundary from the weld metal. Subsequently, the heat affected zone hydrogen concentration decays. If stress is applied soon after welding, critical conditions for cracking are more likely to exist in the weld metal where the hydrogen concentration is high. However, if there is a delay before applying the stress, critical conditions may be achieved in the heat affected zone. Following the contraction model of Satoh and Matsui [15], Matsuda [14] also discussed the role of restraint and showed that with higher restraint levels an increase in stress occurs soon after welding. On the basis of this argument, restraint may be expected to play a part in determining the location of cracking assuming that it is sufficiently high to provide the necessary critical reaction stress.

Present Study

The objective of the present study was to examine the effects of some of the test parameters on the location of cracking in the WIC small scale restraint test. It was thought that, on the basis of the Matsuda model, the level of restraint may affect the location of cracking. In addition, it was thought that preheat might have an effect.

The approach was to analyse the stress and hydrogen distributions in the WIC test using both simple analytical methods and numerical methods and supplement these with a few experimental tests. For the purpose of this study a standard configuration of the test was selected.

The case considered in this study is shown in Fig. 3. Plate thickness is 12 mm and a single root bead having a rectangular cross-section characterizes the weld. The energy input to the weld is taken to be 0.75 kJ/mm, and with a heat efficiency of 80% this provides a heat input of 0.6 kJ/mm.

One-Dimensional Analysis

A one-dimensional analysis has been carried out to estimate the effect of restraint length on the rise of stress and the distribution of hydrogen as a function of time. In the one-dimensional model of contraction first proposed by Satoh and Matsui [15] contraction takes place in two stages: the first soon after welding when the weld is hot and the second some time after when the heat leaves the plates. First stage contraction results from three dimensional heat transfer from the weld into the immediately surrounding plate. If the plate is below a critical thickness, first stage contraction does not occur since heat flow is always two dimensional. Appendix A shows that for the case considered in the present study only second stage contraction occurs.

Second stage contraction results when heat leaves the plate or is conducted beyond the restraint points. These are assumed to remain fixed. For high restraint conditions the restraint points are closer together and the heat reaches them sooner causing second stage contraction to occur early on. Contraction is delayed by reducing the restraint.

The effect of restraint on contraction has been analysed in the present study by considering the heat distribution and consequent thermal expansion in the plates which results in an equal contraction when the heat leaves the plates. Only heat within the restraint length is considered. The analysis is given in Appendix B and leads to the

following expression for the contraction S as a function of time:

$$S = \frac{\alpha H}{g \rho C} \left[1 - \operatorname{erf} \left(\frac{x_0}{\sqrt{4kt}} \right) \right]$$

where α = coefficient of thermal expansion
 H = heat input of weld
 g = plate thickness
 ρ = density
 C = specific heat
 erf = error function
 x_0 = distance from centre of weld to restraint point
 ($x_0 = L/2$ where L is restraint length)
 k = thermal diffusivity
 t = time

It should be noted that in this analysis the initial plate temperature (preheat) does not appear and the contraction is independent of preheat if the specimen is uniformly heated.

Weld contraction for two restraint lengths, 25 and 100 mm, are shown as a function of time in Fig. 4. Increasing the restraint length from 25 to 100 mm delays contraction by a factor of about 10 times. Whereas significant contraction has occurred after 1 minute for a 25 mm restraint length, more than 10 minutes is required to produce the same level of contraction when the restraint length is increased to 100 mm.

The total final contraction is the same in both cases but this would not produce the same final reaction stress because the restraint is different. To determine the reaction stresses an estimate of the intensity of restraint K has been made in Appendix C. Only axial tension and compression between the restraint joints has been considered, and this is bound to overestimate the restraint. A reduction in restraint is afforded by the eccentricity of the weld which allows some bending of the plates. This analysis gave values of $K_{25} = 28,274$ MPa and $K_{100} = 7,048$ MPa for restraint lengths of 25 and 100

mm respectively.

The restraint was also determined using a finite element method (FEM) and gave $K_{25} = 17,364$ and $K_{100} = 8,670$ MPa. Comparison of values suggests that shear in the restraining welds plays a major role under high restraint conditions, and that because of the method of restraining the plates the reduction of restraint due to eccentricity is not as significant as assumed. The values of restraint determined by FEM have been used to estimate the reaction stresses in Appendix C. Due to weld eccentricity the tension stress on the weld face is higher than that on the root, and this may play a major role in determining the location of crack initiation.

The stress as a function of time is shown plotted in Fig. 5 for the two restraint lengths of 25 and 100 mm. Final stress is lower with the 100 mm restraint length as expected, and the rise of stress is considerably delayed.

The stresses shown in Fig. 5 are mean stresses across the weld face and do not include the effect of stress concentrations at the weld toe. These may typically raise the stress four times. The values of elastic stress calculated have indicated that yielding is expected and that significant plastic straining of the weld metal will occur.

Hydrogen Diffusion

Estimates of the distribution of hydrogen in the weld metal and heat affected zone were made by assuming simple one dimensional diffusion from a slab of half thickness l . For the weld metal, hydrogen is assumed to be lost from the face and the root (Fig. 6a) and setting the concentration equal to zero at these two surfaces defines the boundary conditions.

Because the temperature and diffusion coefficient D is constantly changing, integrated values of $D\Delta t$ have been used. Values were taken from previous work [11] on G-BOP tests which used a similar size weld bead. Diffusion above 500°C was neglected. The level of hydrogen at a point 0.36 mm (0.2*l*) below the weld face was determined for preheats of 20°C and 100°C .

HAZ hydrogen levels close to the fusion boundary were calculated assuming one-dimensional diffusion from the weld metal (Fig. 6b). The weld metal is assumed not to lose hydrogen from the faces. Over the long term this assumption is, of course, not true but since the diffusion distances into the HAZ are quite small, the assumption is reasonable for the initial period until the HAZ hydrogen level reaches that of the weld metal. Thereafter it will decay as the overall hydrogen level decreases.

The results of these calculations are shown in Figs. 7 and 8 for the two preheats. With the higher preheat the weld metal hydrogen decays more rapidly, and the level in the HAZ reaches a peak earlier. However, about an hour after welding, when the assembly has cooled, the weld metal and HAZ have similar hydrogen levels for a 100°C preheat. For the 20°C preheat the weld metal hydrogen is substantially more than that of the HAZ after about an hour.

These results suggest that low, or no preheats might favour weld metal cracking whereas higher preheats might encourage HAZ cracking.

Finite Element Analysis

Analysis of stress and hydrogen distributions was carried out by Fleet Technology Limited using a Finite Element Method (FEM). Full details of this phase of the work are contained in Appendix D as a separate report. The results show a pattern for the decay of weld metal hydrogen content (Appendix D, Figs. 19 and 22) that is similar to that estimated from the simple 1-D analysis. The FEM analysis gives a slightly faster decay rate.

Figs. 26 and 33 of Appendix D show the increase of stress on the weld face as a function of time. A lower final stress results from the lower restraint, but these graphs do not reveal a significant delay at the lower restraint. Further comparison with Figs. 44 and 47 of Appendix D indicates very little effect of preheat on the stress which is in agreement with the 1-D model.

Stresses on the lower (root) face of the weld are lower than on the top surface which agrees with the idea that weld eccentricity introduces an additional bending moment. Final stresses for the RT preheat determined by the FEM and the simple 1-D analysis are compared in Table 1. Agreement is not unreasonable recognizing the assumptions made in the simple model. The FEM analysis took into account yielding of the weld metal, and this explains the large difference in the values on the weld face for the 25 mm restraint length. It also appears that the 1-D model over-estimates the effect of weld eccentricity.

Experimental Tests

Six WIC small scale restraint tests were carried out at the Welding Institute of Canada using conditions suggested by the results of the 1-D and FEM analysis. The steel and electrode were chosen such that the HAZ and weld metal would have similar hardnesses and neither would dominate in terms of microstructural susceptibility to hydrogen cracking. The composition of the steel is given in Table 2 and Table 3 shows the welding conditions employed. The six tests were designed to cover three levels of restraint and two preheats. The 60°C preheat was selected on the basis of previous experience to provide a small degree of cracking which, hopefully, would allow identification of the crack initiation point. However, since no significant cracking occurred in these two tests, a further test was conducted at a preheat of 45°C.

Table 4 shows the actual welding conditions achieved in the six tests and indicates that heat input levels were reasonably consistent. The amount of cracking observed is shown in Table 5. Three sections were examined in each weld, and in all room temperature and the 45°C tests the sections were completely cracked. The two tests at 60°C showed no significant cracking, one section only having a very small weld metal microcrack.

Photographs of two sections from each of the room temperature welds are shown in Figs. 9 to 14. All three welds show cracking in both the HAZ and the weld metal, and based on the width of the cracks it is probable that initiation occurred in the weld metal toe. In the highest restraint (25 mm) test one section (Fig. 9) showed the crack in the weld metal centre line perhaps suggesting a greater susceptibility at this location.

However, one of the sections (Fig. 14) for the lowest restraint test also showed 100% cracking in the weld metal. The results taken together do not suggest any significant shift of cracking location due to changes in restraint. The microcrack in one of the 60°C preheat tests is shown in Fig. 15 and appears to be due to a local hard spot.

Hardness measurements taken on transverse sections of the welds made with room temperature and 60°C preheats gave the results given in Figure 16. Hardness levels are marginally lower in the preheated sample, but the HAZ and weld metal have similar values. This suggests that hardness is not a major factor in this case in determining the location of cracking.

Discussion

Many factors could influence the location of cracks in cracking tests including inherent susceptibility of the microstructure, geometry, and hydrogen distribution. The present study has looked at stress and hydrogen distribution as a function of test parameters to determine whether these relate to the location of cracking.

The simple one-dimensional model of contraction suggests that decreasing the restraint causes a delay in the development of the reaction stress which could promote HAZ cracking. This delay, however, was not evident in the FEM analysis even though final reaction stresses were of a similar magnitude. The reason for this discrepancy is not clear since the expectation of a delay is clearly supported by the evidence of rigid restraint cracking (RRC) tests [15]. The 1-D model neglected any surface transfer of heat, and if this were large it would tend to diminish any delay between the high and low restraint reaction stresses. The rapid rise of stress with a 100 mm restraint length predicted by the FEM analysis implies that most of the heat in the plates within the restraint length has been dissipated within the first minute after welding which seems unlikely.

Both the FEM and the 1-D model predict a difference between the surface stress on the top and the bottom of the weld due to eccentricity, and this is believed to play a major role in crack location as others [16] have pointed out. In the present case, initiation on the top face at the toe would be favoured, and this appears to be supported

by the experimental evidence. Initiation under these conditions is therefore more likely to be in the weld metal for geometrical reasons.

If the weld were placed at the midthickness or in the top half of the thickness, the eccentricity would increase tension in the root and favour HAZ initiation because of the local geometry. This aspect of geometry has yet to be systematically studied in the WIC test.

The hydrogen diffusion models used in the present work did not consider any effects of phase differences across the fusion boundary. Thus hydrogen concentration is continuous across the boundary and within a short period after welding the levels are equal for a short distance either side of the boundary. On this basis there is no reason to favour initiation in the HAZ or the weld metal on account of the hydrogen concentration. It is true that the models predict a higher hydrogen in the centre of the weld and if stress is applied early this might be a favoured place for crack initiation.

One of the sections for the 25 mm restraint length test did, in fact, show cracking in the weld metal centre thus supporting this idea. In Matsuda's work the centre was also observed to be the crack location when stress was applied early.

A more significant effect on local hydrogen distribution may be that due to differences in transformation temperature. Diffusion rates in austenite are very low and little hydrogen escapes from the weld until it transforms to a ferritic microstructure and the diffusion rate increases. If this does not occur until a low temperature, little opportunity remains to lose hydrogen, and the resulting level is high. Thus the retained hydrogen level in the weld metal is expected to relate directly to the transformation behaviour. This effect may be very significant in high strength alloyed weld metals and warrants further investigation.

Transformation temperature may have a further important effect at the fusion boundary. If a phase difference exists at the boundary the differences in hydrogen activity will cause a local 'pumping' of the hydrogen from the ferritic to the austenitic side causing a short range depletion in the ferrite and accumulation in the austenite. If this difference persists to low temperatures it may result in favouring cracking in the

last material to transform. The effectiveness of austenitic electrodes in welding high carbon steels with martensitic HAZ's probably results from this phenomenon. A detailed study of this effect of local hydrogen distribution was beyond the scope of the present study but should be considered as a factor in determining the preferred crack initiation site.

The experimental results in the present study, although limited, do not support the idea that restraint length can be used to control the location of cracking. Neither does the preheat appear to affect the location although there were insufficient results to draw a firm conclusion. Both the 1-D and FEM analysis showed little effect of preheat.

Conclusions

The effects of the restraint length and preheat on the location of cracking in the WIC small scale restraint test have been investigated using analytical methods supplemented by limited experiments. The results lead to the following conclusions:

1. A simple one-dimensional analysis predicts a significant delay in the development of the reaction stress when the restraint length is increased from 25 mm to 100 mm. The FEM analysis does not predict such a delay.
2. Both analytical methods predict higher stresses on the face (top) of the weld than the root (bottom) due to bending moments arising from the weld eccentricity.
3. The hydrogen level in the heat affected zone at a distance from the fusion boundary reaches a peak after welding before declining. Increasing preheat causes the peak to occur earlier. However, hydrogen levels within a short distance of either side of the fusion boundary are similar when effects of phase transformations are neglected.
4. The limited experiments showed no effect of restraint length on the location of cracks.

It is recommended that further work should be done to determine why the 1-D and FEM analyses predict different effects of restraint on the delay of stress development. Further studies should also investigate the effects of geometry and phase transformation. A systematic study of the effects of joint preparation and eccentricity

should be carried out using FEM modelling supplemented with experimental testing. The effects of transformation temperatures on local hydrogen distribution should consider both the effects of weld metal transformation on retained hydrogen levels and the effects of differences in transformation temperature between the HAZ and the weld metal on local hydrogen redistribution across the fusion boundary.

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Table 1. Mean stress across faces of weld (MPa)

	25 mm restraint length		100 mm restraint length	
	Face	Root	Face	Root
FEM	750	550	575	350
1- D	1030	423	500	205

Table 2. Composition of steel used in restraint tests (wt %).

Thickness = 12 mm Pipe diameter = 1067 mm.

Source: IPSCO heat 434337

C	0.06
Mn	1.59
S	0.005
P	0.014
Si	0.18
Cu	0.33
Ni	0.14
Cr	0.12
V	0.002
Nb	0.065
Mo	0.188
Sn	0.014
Al	0.075
N	0.013
Ti	0.036
Ca	0.004

Table 3. Welding conditions selected for the restraint tests.

Current:	120 A
Voltage:	26 V
Speed:	4.17 mm/s
Energy input:	0.75 kJ/mm
Electrode:	E8010G
Groove angle:	60~65°
Root opening:	1.6 mm
Root face:	1.6 mm

Table 4. Actual welding conditions achieved in the restraint tests.

Test #	Restraint length (mm)	Preheat °C	Current (amps)	Time for 50 mm (s)	Energy input kJ/mm
1	25	RT	125	11.2	0.73
2	75	RT	122	8.5	0.54
3	100	RT	126	11.4	0.75
4	25	60	126	11.9	0.78
5	75	60	127	12.5	0.83
6	75	45	120	11.3	0.71

Table 5. Results of restraint tests.

Test #	Restraint length (mm)	Preheat °C	Section #	Crack length (mm)	Crack ratio %
1	25	RT	1	T	119*
			2	T	
			3	T	
2	75	RT	1	T	100
			2	T	
			3	T	
3	100	RT	1	T	100
			2	T	
			3	T	
4	25	60	1	-	0
			2	-	
			3	-	
5	75	60	1	0.08	0.6
			2	-	
			3	-	
6	75	45	1	T	100
			2	T	
			3	T	

T = Total section cracked

* Centreline crack

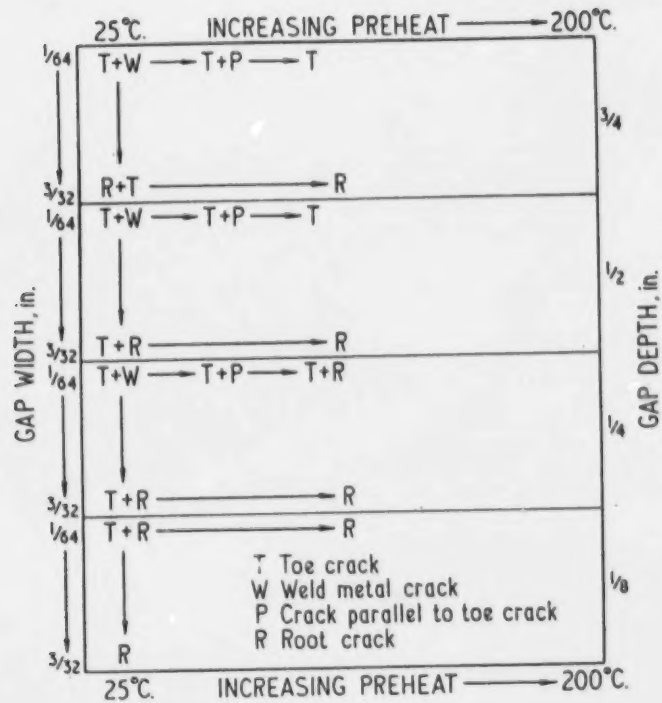


Figure 1. Classification of crack location in CTS tests on QT45 steel (reference 5)

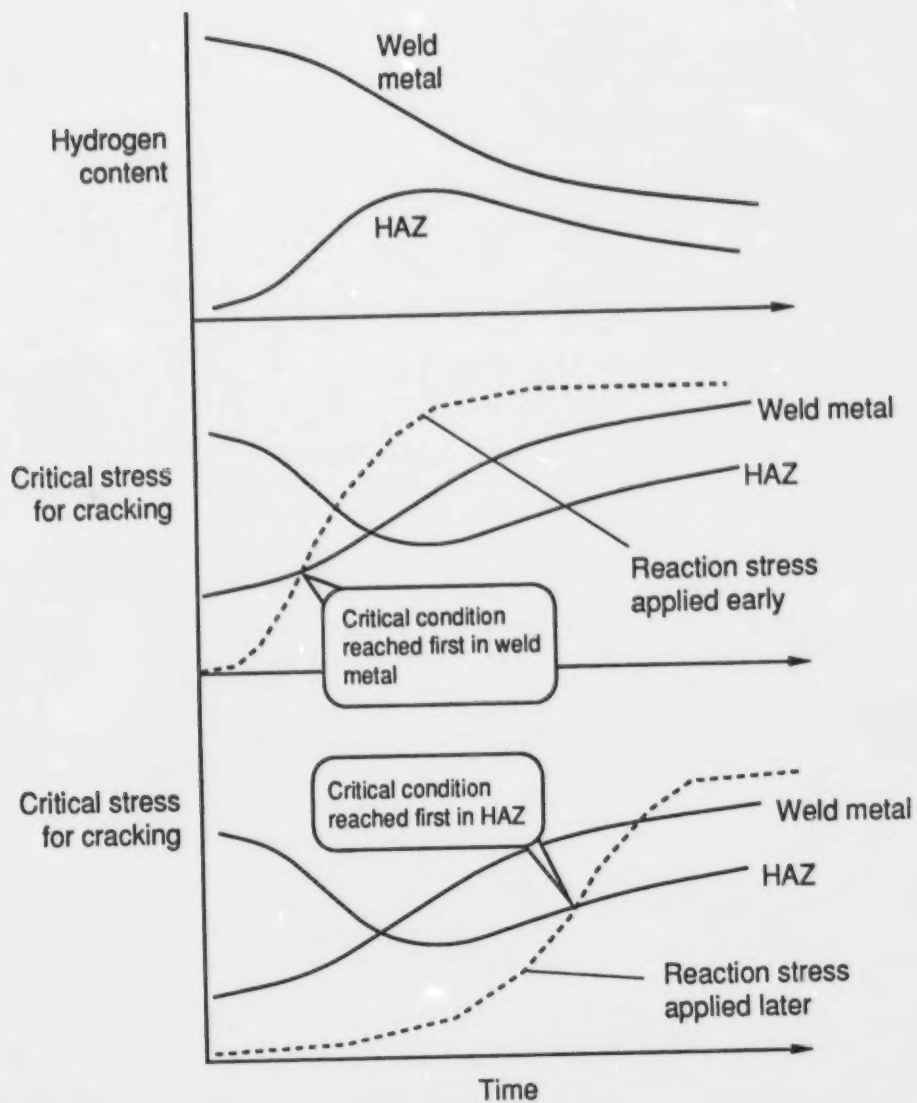


Figure 2. Matsuda model for criterion for crack location.

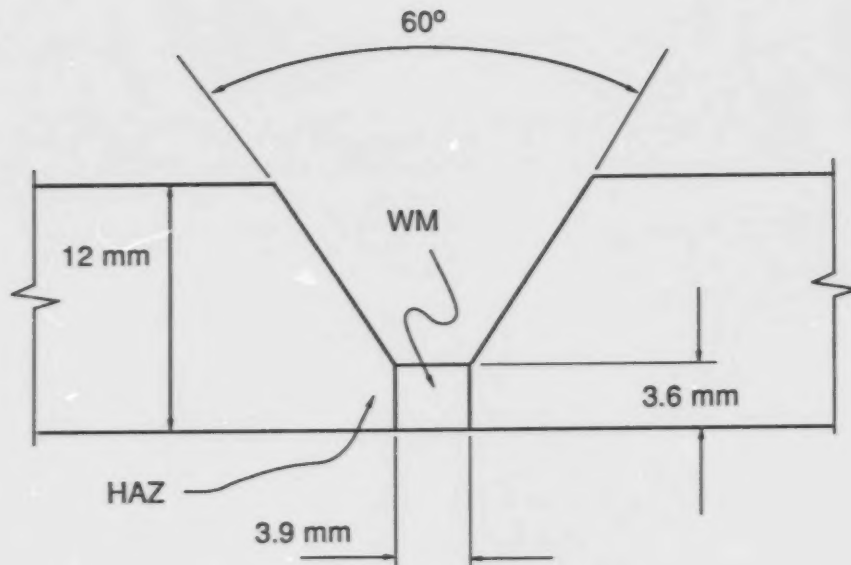


Figure 3. Geometry of weld considered in the present study.

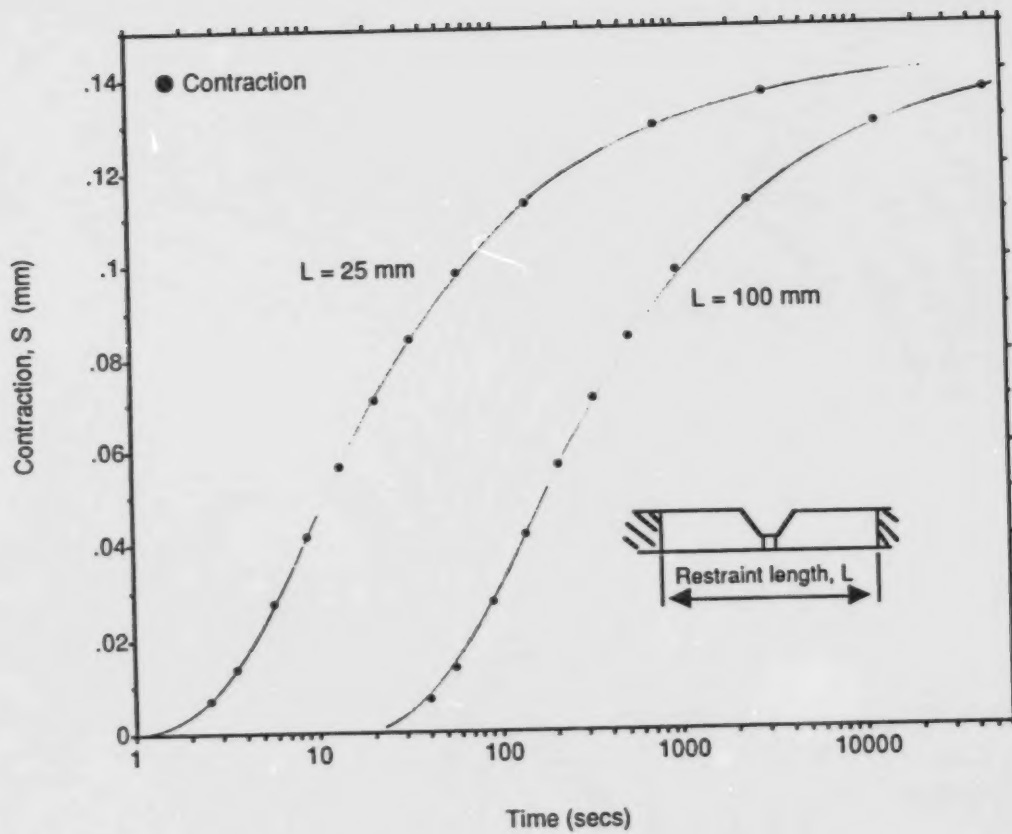


Figure 4. Contraction S as a function of time for two restraint lengths as determined from the one-dimensional analysis.

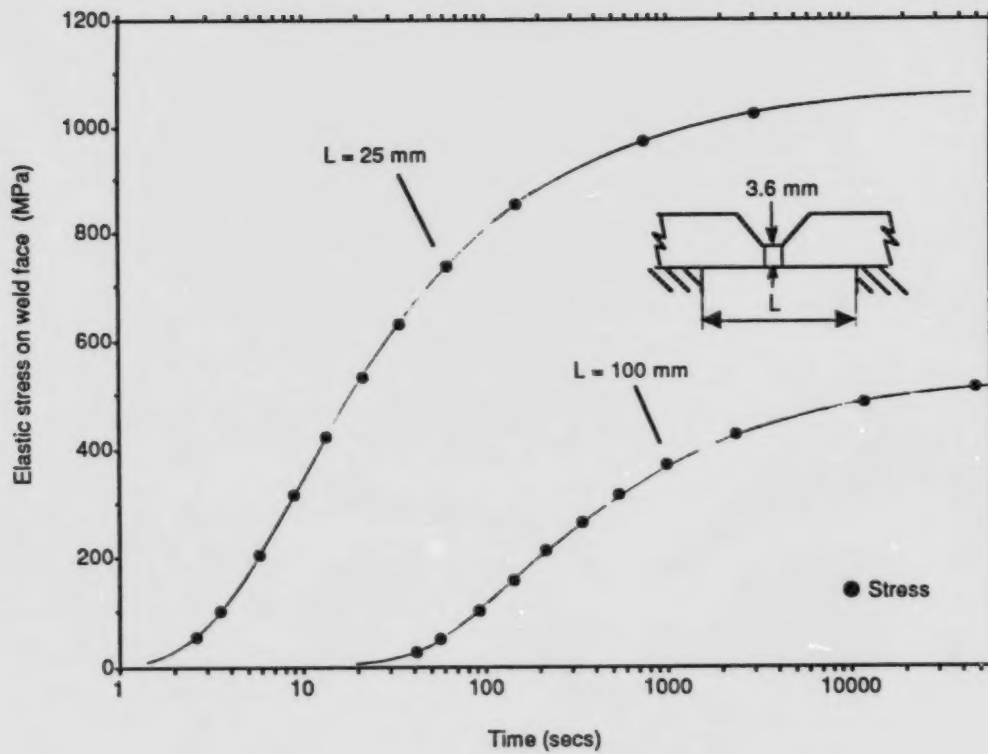


Figure 5. Reaction stress as a function of time for two restraint lengths as determined from the one-dimensional analysis.

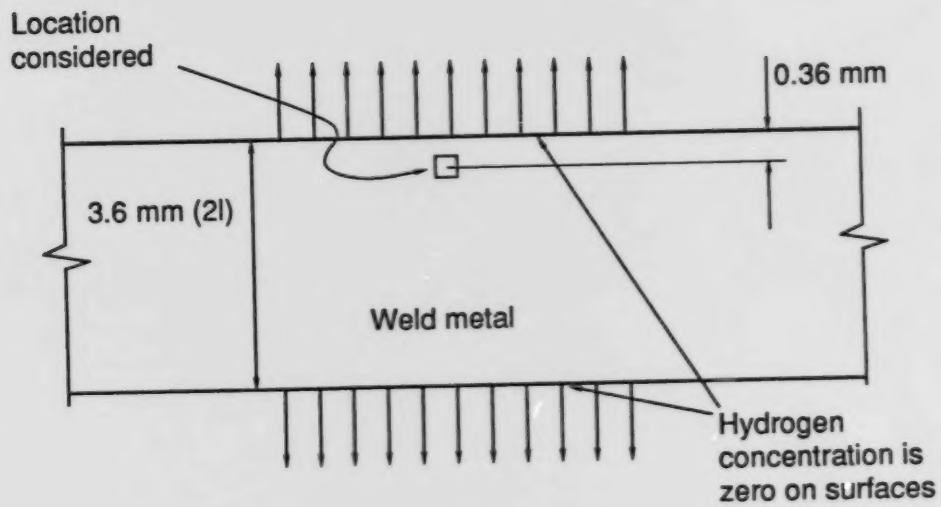


Figure 6a. Diffusion model for weld metal.

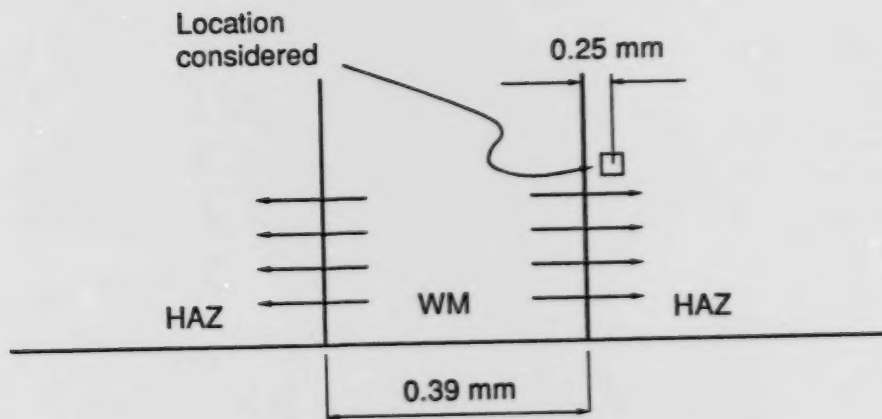


Figure 6b. Diffusion model for heat affected zone.

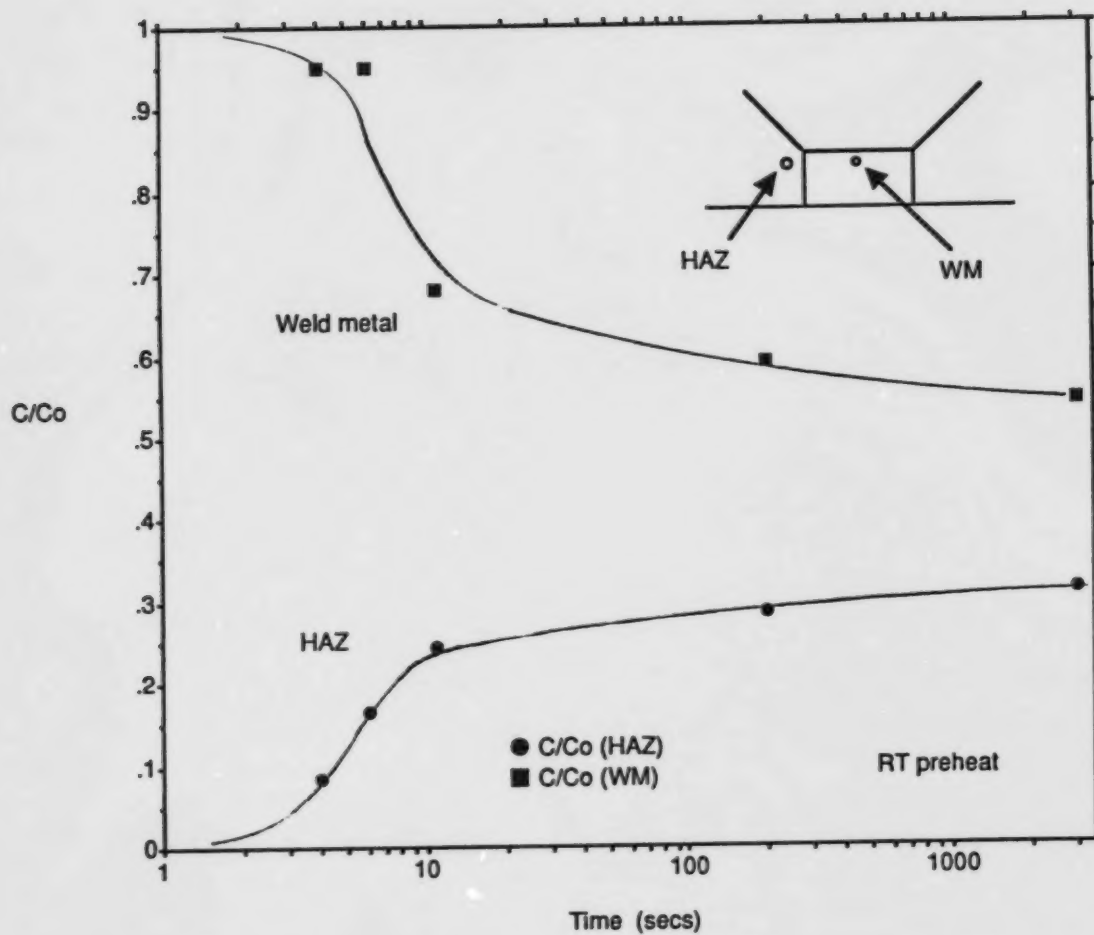


Figure 7. Distribution of hydrogen from simple one dimensional model (no preheat).

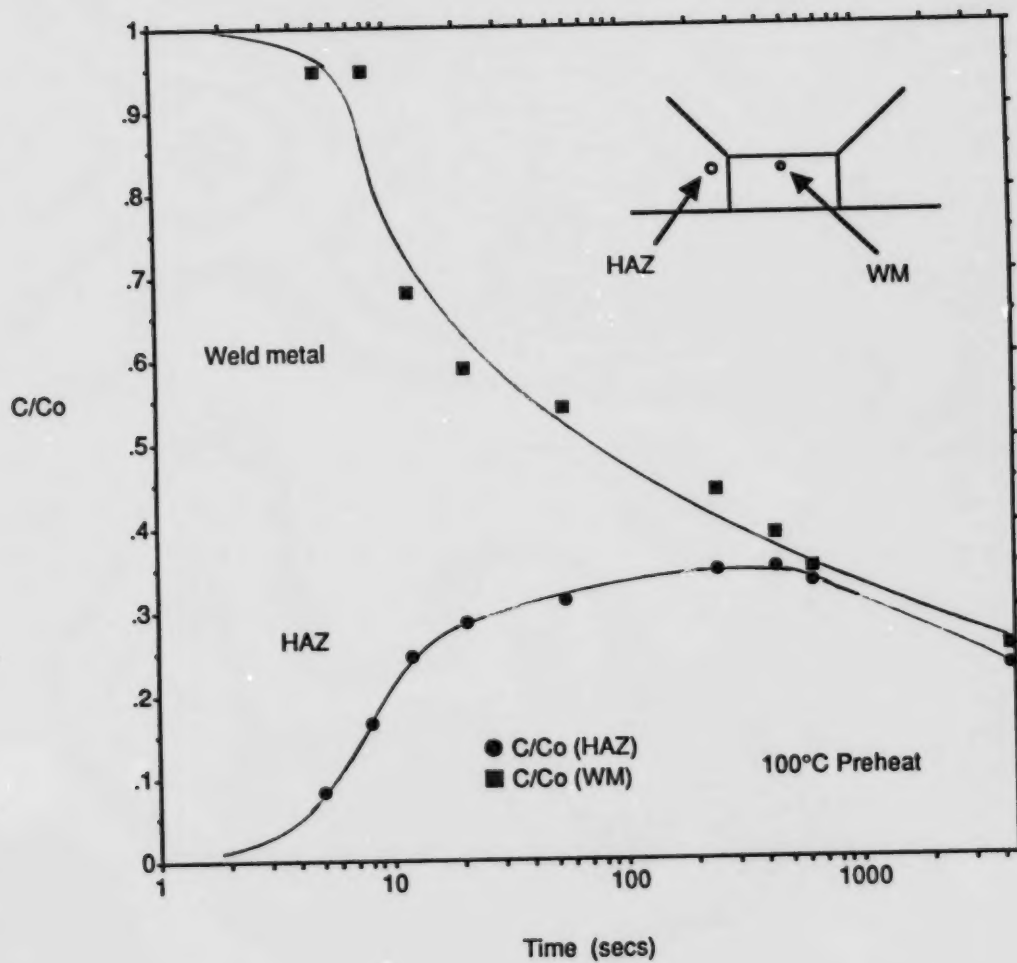


Figure 8. Distribution of hydrogen from simple one-dimensional model (100°C preheat).

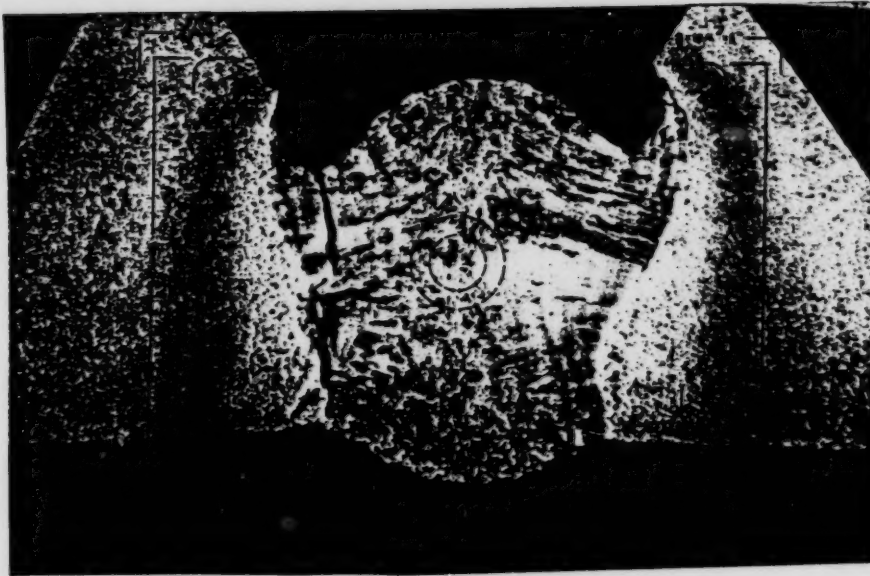


Figure 9. Section from weld #1 (restraint length = 25 mm)

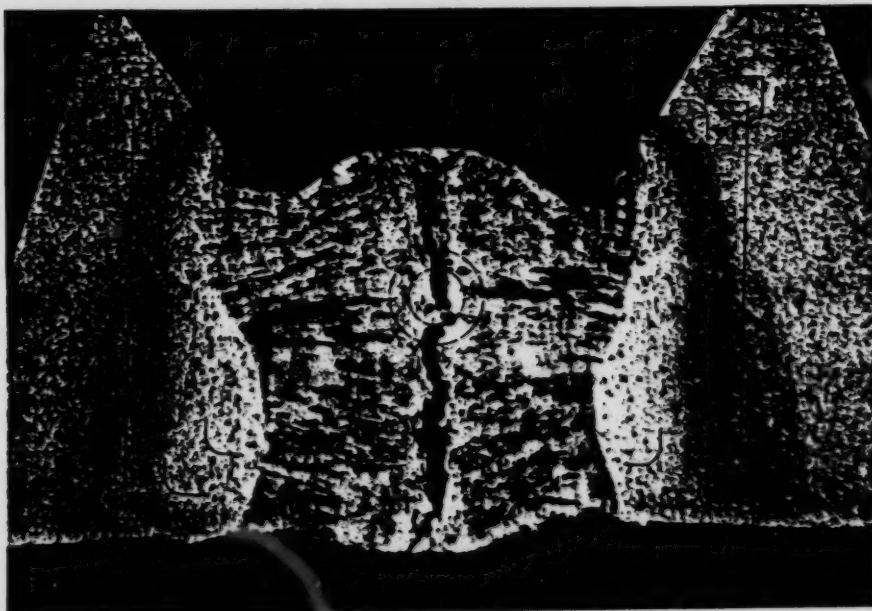


Figure 10. Section from weld #1 (restraint length = 25 mm)

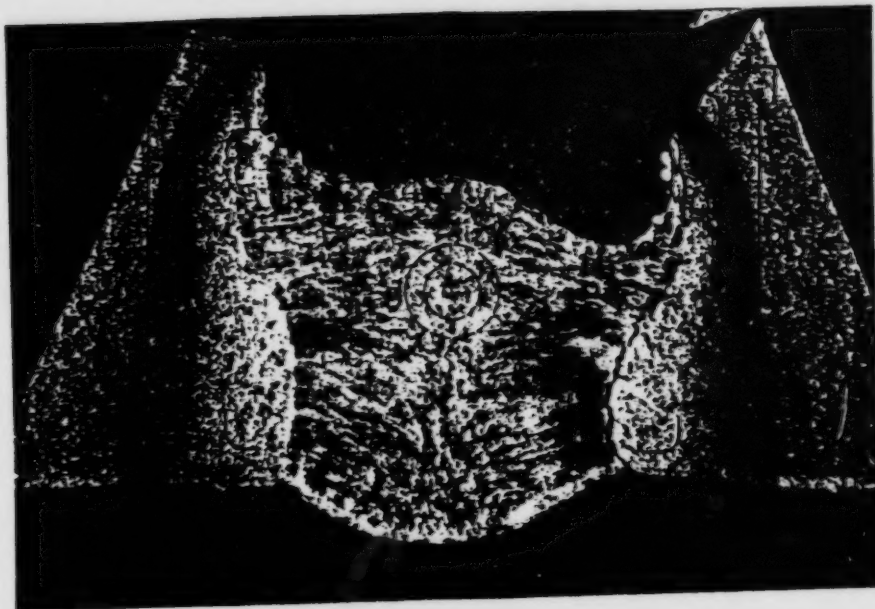


Figure 11. Section from weld #2 (restraint length = 75 mm).

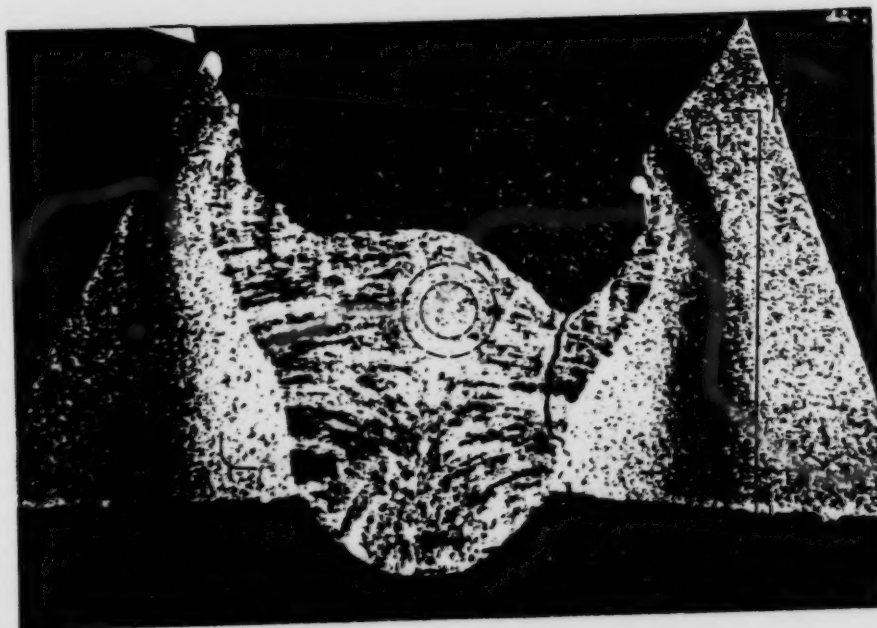


Figure 12. Section from weld #2 (restraint length = 75 mm).

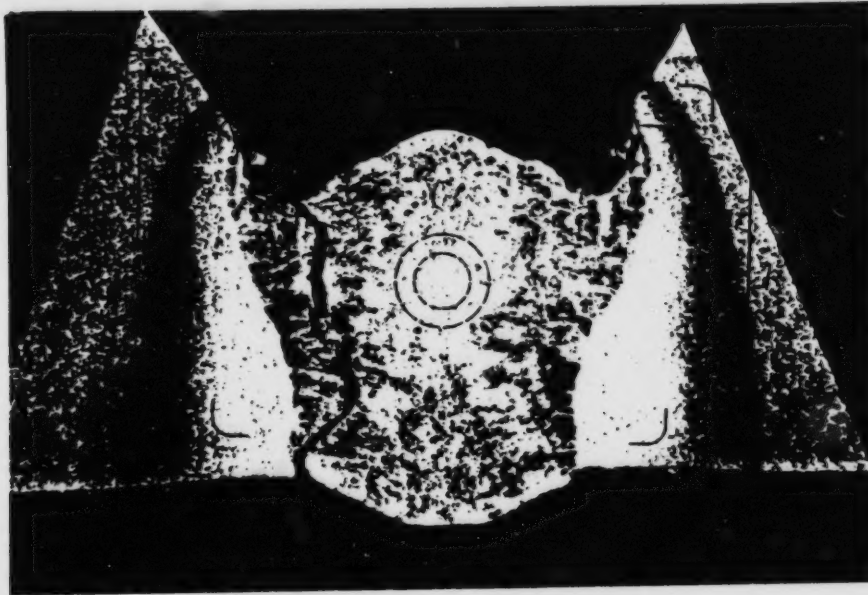


Figure 13. Section from weld #3 (restraint length = 100 mm).

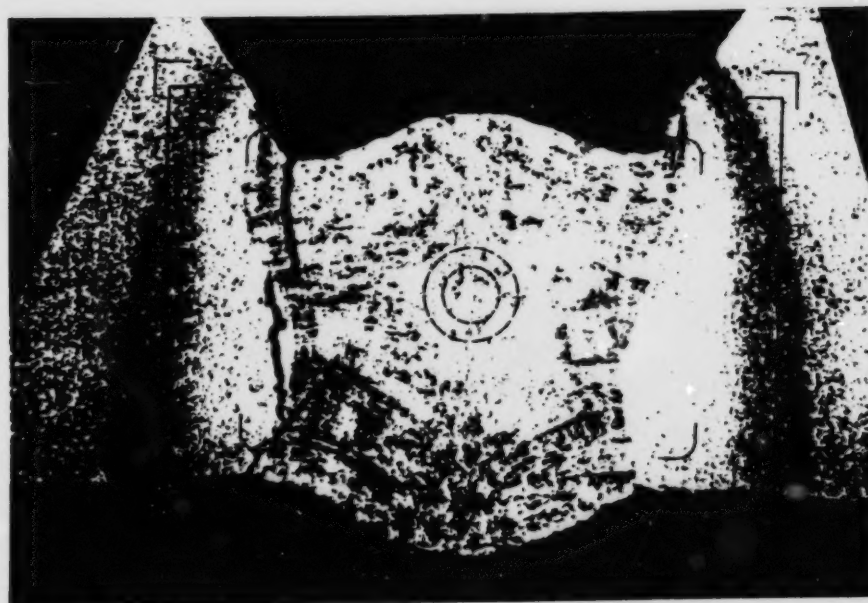


Figure 14. Section from weld #3 (restraint length = 100 mm).

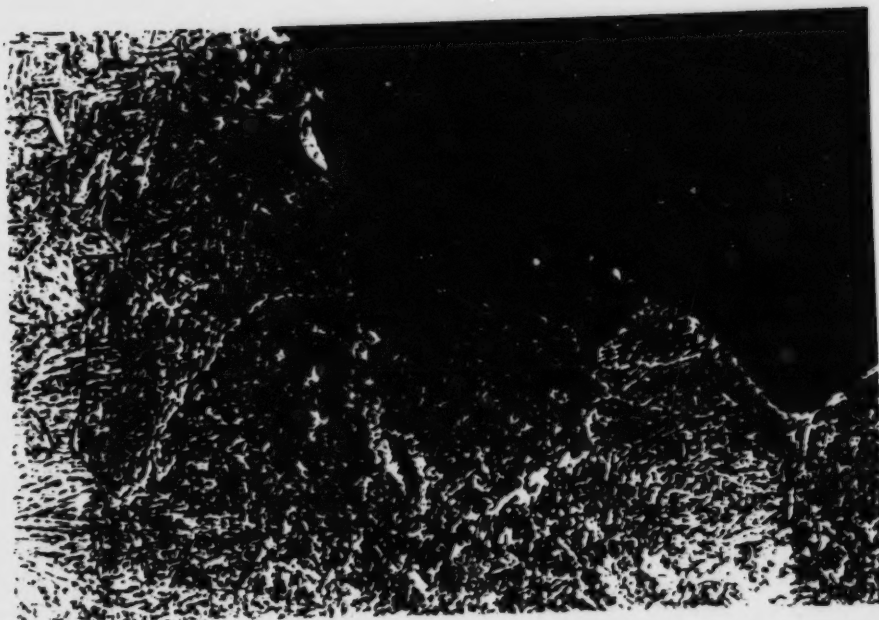
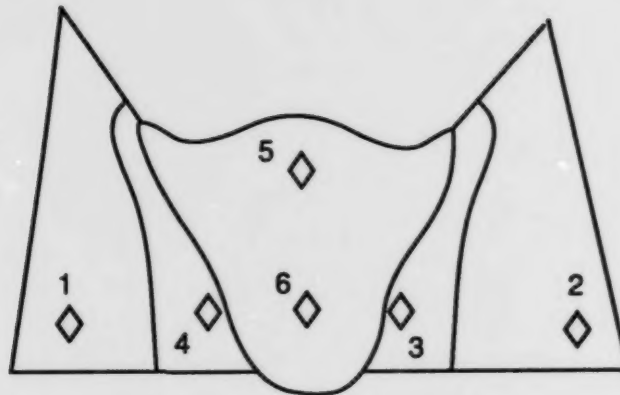


Figure 15. Microcrack in the weld metal of test # 5.



No preheat, restraint length = 25 mm

Location	Vicker's hardness (Hv 5 kg)
1 Parent	246
2 Parent	249
3 Coarse grained HAZ	313
4 Coarse grained HAZ	310
5 Weld metal	296
6 Weld metal	296

60°C preheat, restraint length = 25 mm

Location	Vicker's hardness (Hv 5 kg)
1 Parent	229
2 Parent	229
3 Coarse grained HAZ	293
4 Coarse grained HAZ	289
5 Weld metal	296
6 Weld metal	280

Figure 16. Hardness results on sections of welds from restraint tests.

APPENDIX A

Critical Thickness

It is assumed that contraction occurring above 800°C does not contribute to reaction stresses because the weld metal yield strength is too low. The critical thickness below which all contraction is second stage is

$$g_c = \left(\frac{H}{\rho C (800 - 20)} \right)^{1/2}$$

where H = heat input = 0.8×0.75 kJ/mm
 ρ = density
 C = specific heat
20 = initial plate temperature or preheat.

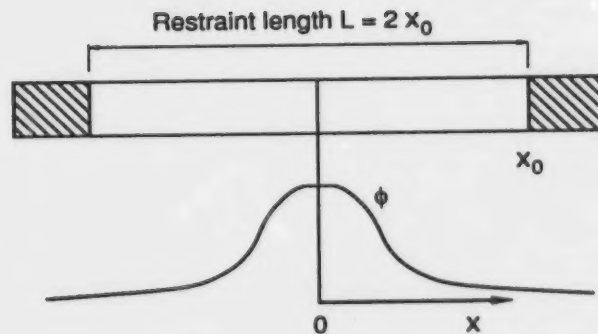
Inserting appropriate values gives

$$g_c = 11.5 \text{ mm} \sim 12 \text{ mm (thickness being considered)}$$

For a preheat higher than 20°C, g_c is greater. It is thus concluded that for the 12 mm plate and bead size considered the total contraction is all second stage.

APPENDIX B

Contraction



It is assumed that the net contraction is that due to contraction of the weld and plate within the restraint length. The restraint points are therefore assumed to be fixed.

The total free contraction (or expansion) E_T of the plate is given by

$$E_T = \frac{\alpha H}{g \rho C}$$

where α is the coefficient of expansion and is assumed to be independent of temperature.

The expansion of the unrestrained length at time t is

$$E = 2\alpha \int_0^{x_0} \phi(x, t) dx$$

where $\phi(x, t)$ is the temperature above the initial plate temperature.

The net contraction S that has taken place up to time t is therefore:

$$S = E_T - E$$

$$= \frac{\alpha H}{g \rho C} - 2\alpha \int_0^{x_0} \phi(x, t) dx$$

Since all second stage contraction occurs when heat flow is linear, the temperature distribution can be established from the delta function solution (line heat source):

$$\phi(x, t) = \frac{H}{g \rho C \sqrt{4\pi k t}} \cdot \exp\left[-\frac{x^2}{4kt}\right]$$

where k = thermal diffusivity.

Therefore:

$$S = \frac{\alpha H}{g \rho C} - \frac{2\alpha H}{g \rho C \sqrt{4\pi k t}} \cdot \int_0^{x_0} \exp\left[-\frac{x^2}{4kt}\right] dx$$

Making the substitution $\beta^2 = \frac{x^2}{4kt}$ allows evaluation of the integral:

$$\int_0^{x_0} \exp\left[-\frac{x^2}{4kt}\right] dx = \sqrt{4kt} \int_0^{\frac{x_0}{\sqrt{4kt}}} \exp[-\beta^2] d\beta = \sqrt{\pi k t} \operatorname{erf} \frac{x_0}{\sqrt{4kt}}$$

where erf is the error function.

Therefore:

$$S = \frac{\alpha H}{g \rho C} \left[1 - \operatorname{erf}\left(\frac{x_0}{\sqrt{4kt}}\right) \right]$$

APPENDIX C

Restraint

A conservative (high) estimate of the intensity of restraint is made by assuming elastic extension of the test plates and elastic compression of the restraint plates between the anchor points. This neglects shear and bending of the plates. If A is the cross-sectional area of the restraining plates, L the restraint length, and B the weld length then the restraint is

$$K = \frac{\frac{AEg}{LB}}{\frac{A}{B} + g} = \frac{Eg}{L \left(1 + \frac{gB}{A} \right)}$$

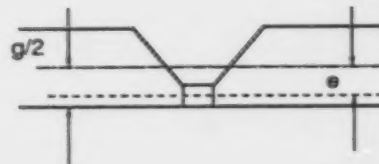
Inserting the values:

$$\begin{aligned} A &= 4000 \text{ mm}^2 \\ B &= 50 \text{ mm} \\ g &= 12 \text{ mm} \\ L &= 25 \text{ or } 100 \text{ mm} \\ E &= 21 \times 10^4 \text{ MPa} \\ K &= \text{restraint} \end{aligned}$$

$$K_{25} = 8.765 \times 10^4 \text{ MPa}$$

$$K_{100} = 2.191 \times 10^4 \text{ MPa}$$

These values will be reduced because of eccentricity of the weld which allows some bending of the test plates. Following Suzuki [17] and Satoh [15] the restraint



under eccentric conditions is

$$K_e = \frac{K}{1 + 3\gamma^2} \text{ where } \gamma = \frac{e}{g/2} = \frac{4.2}{6} = 0.7$$

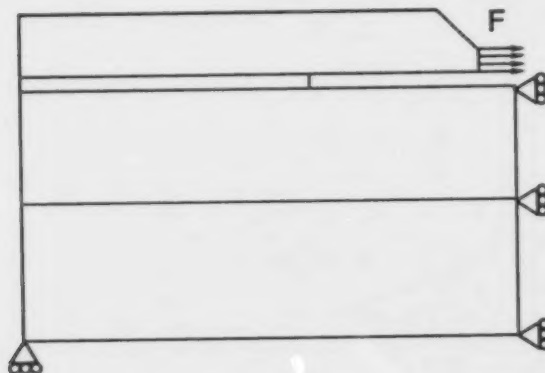
$$\therefore K_e = \frac{K}{3.1}$$

Therefore the corrected values of restraint for 25 and 100 mm restraint lengths are:

$$K_{25} = 28,274 \text{ MPa}$$

$$K_{100} = 7,048 \text{ MPa}$$

These are compared with estimates made by FEM analysis using the boundary conditions shown below



FEM estimates:

$$K_{25} = 17,364 \text{ MPa}$$

$$K_{100} = 8,670 \text{ MPa}$$

The FEM values are used below.

Reaction Stress

Under elastic conditions the mean weld metal final reaction stress is

$$\sigma = \frac{KS}{g_w} \text{ where } g_w \text{ is the thickness of the weld metal.}$$

$$\sigma_{25} = \frac{17,364 \times 0.142}{3.6} = 685 \text{ MPa}$$

$$\sigma_{100} = \frac{8,670 \times 0.142}{3.6} = 342 \text{ MPa}$$

These values will be further modified by bending of the weld metal due to eccentricity and local stress concentrations. Bending causes the tension stress at the weld root to be reduced and that on the face (where cracking initiates) to be increased. The increase is given by Suzuki [17]

$$= (1 + B\gamma) \text{ where } B = 0.8 \text{ and } \gamma = 0.7$$

$$= 1.56$$

$$\therefore \text{ Elastic stress on face } \sigma_{F(25)} = 685 \times 1.56 = 1,069 \text{ MPa}$$

$$\sigma_{F(100)} = 342 \times 1.56 = 534 \text{ MPa}$$

Local stress concentrations at the weld toe could increase these values by about 4 times.

The increase of reaction stress on the weld face as a function of contraction is therefore given by

$$\sigma_{F(25)} = \frac{1,069}{0.142} \times S = 7,528 \times S \text{ MPa}$$

$$\sigma_{F(100)} = \frac{534}{0.142} \times S = 3,761 \times S \text{ MPa}$$

where 0.142 is the total final contraction.

These values were used to generate the curves in Fig. 5.

APPENDIX D

**NUMERICAL MODELLING OF WELD ZONE STRESSES AND
HYDROGEN DIFFUSION**

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SUMMARY

Numerical (finite element) modelling of standard WIC test has been performed involving two preheat levels (20°C and 100°C) and two restraint lengths (25mm and 100mm). Results of thermal analysis, hydrogen diffusion analysis and stress analysis are presented in this report.

1.0 INTRODUCTION

Hydrogen induced cold cracking is always one of the major considerations in designing welding procedures for steels. Often these cracks are formed at temperatures below 200°C, and may sometimes exhibit a delay of several days.

Generally speaking, cracks including cold cracks are not tolerated in a structure. Since they may sometimes escape detection and are expensive to repair, it is essential that proper precautions are taken during fabrication to prevent their occurrence.

Based on studies in the past, it is now recognized that hydrogen cold cracks are due to the interaction of the following three factors:

- (a) the hydrogen content in the weld metal;
- (b) stress levels introduced as a result of welding; and
- (c) the temperature-time cycle.

Under certain situations, one of the factors may dominate.

In order to predict whether cold cracking will occur during fabrication, various tests have been devised to study the cracking phenomenon. One of the tests, "Small Scale Restraint Test", is used at the Welding Institute of Canada, and is shown in Figure 1. This test essentially consists of a 12mm test plate with a 60° groove joint, restrained on top of a Tee section using fillet welds. The test plate sits on a shim, so that essentially heat loss to the Tee section is small during welding. By changing the length of the fillet weld, the restraint length (l), and hence the restraint of the joint can be varied. Any changes in the restraint of the joint will affect the stress distribution, and hence the incidence and perhaps the location of cracking.

Since diffusivity of hydrogen in ferritic lattice decreases rapidly with temperature, changes in the preheat conditions of the test plate, will affect the movement of hydrogen. This in turn will affect the hydrogen cold cracking. The hydrogen diffusivity as a function of temperature as given in [1], and used as input data in this program, is shown in Figure 2.

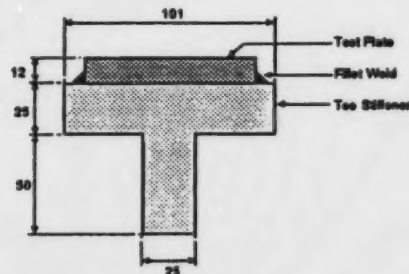
The present project was initiated in order to numerically study the effect of restraint and preheat conditions on hydrogen diffusion and stresses developed in the Small Scale Restraint Test.

2.0 CONDITIONS FOR FINITE ELEMENT (FEM) ANALYSIS

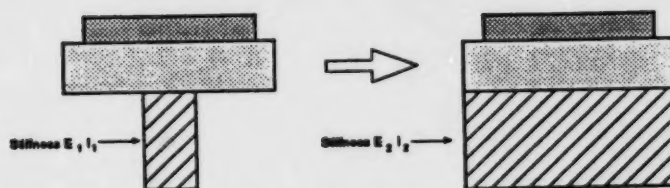
- Geometry : Standard WIC Test (Figure 1)
- Test plate thickness : 12mm
- Welding Conditions : 120A, 26V, 4.17 mm/s (heat input=0.75 kJ/mm)
- Welding Process : SMAW (weld efficiency assumed = 0.70)
- Weld Metal Hydrogen Content : 50 ml/100gm of fused metal
- Mechanical Properties : Yield strengths of weld metal, HAZ and base plate assumed are as shown in Figure 3. Other mechanical properties needed (such as Poisson's ratio, coefficient of thermal expansion, Young's Modulus as functions of temperature) were obtained from handbooks.
- Thermal Properties : Thermal conductivity and volumetric specific heat values as function of temperature used in the analysis are shown in Figures 4 and 5.
- Hydrogen Diffusivity : Hydrogen diffusivity values as a function of temperature used in the analysis are shown in Figure 2. Transformation temperature for weld metal and HAZ assumed to be 600°C.
- Model Matrix: : (i) Two preheats - Room Temperature (20°C) and 100°C
- (ii) Two restraint lengths - 25mm and 100mm

3.0 FINITE ELEMENT SOLUTION (FEM)

The FEM solution involved different steps. The first step was to generate the FEM mesh representation of the actual geometry. Since the present analysis was 2 dimensional, the stiffness of the FEM geometry should represent the stiffness of the test set-up. For this, the following dimensions of the test set-up were considered:



Since the web of the tee could not be represented adequately in 2D, an equivalent modulus for the web was calculated, such that the stiffness of FEM geometry and test geometry were equivalent. This was done to ensure that the neutral axis of both the set-ups were the same.



Stiffness $E_1 I_1$

$$\text{i.e. } E_1 \times \frac{1}{12} \times 25 \times 50^3$$

Stiffness $E_2 I$

$$E_2 \times \frac{1}{12} \times 100 \times 50^3$$

$$\text{so, } E_2 = \frac{E_1}{4}$$

The generated FEM mesh representing the test set-up is shown in Figure 6. Since the test set-up had a plane of symmetry, only one half of the cross-section was modelled for analysis.

The second step in the analysis was to perform the two thermal analyses at preheats of 20°C and 100°C using the welding conditions of 120A, 26V and 4.17 mm/s travel speed. The weld heat was represented by a double ellipsoidal Gaussian distribution [2]. A convective heat transfer coefficient of $2.4 \times 10^{-5} \text{ W/mm}^2$ was assumed on the top surface (Figure 7).

The transient thermal history provided the necessary thermal load for an elasto-thermo plastic stress analysis. Four stress analyses were performed corresponding to two preheats (20°C and 100°C) and two restraint lengths (25mm and 100mm). The peak stresses in the analyses are limited by the yield surface.

The next step was to perform hydrogen diffusion analysis using the transient temperature distribution as the driving force. To do this, the temperature-time curves were divided into small steps and diffusion analysis performed.

For hydrogen diffusion, it was assumed that the medium is homogeneous, and as such, Fick's Law is applicable and diffusion coefficient is a function of temperature only. Because of strain fields, voids and traps, etc., this is not strictly true. Japanese researchers[3] incorporate activity coefficients, diffusivity as functions of percentage strain etc. in their analyses but the detailed approach is not clear.

In order to see the effect of transformation temperature, two additional hydrogen diffusion analyses were performed for the room temperature preheat condition (Appendix A). In one case, the weld metal transformation was assumed to be at 500°C, and in the second case, it was 700°C. Heat affected zone transformation temperature was 600°C as in the first four runs.

4.0 RESULTS

4.1 Thermal Analysis

Peak temperature isotherms for the two preheats are shown in Figures 8 and 9. The isotherm representing 1500°C can be regarded as the fusion boundary.

The temperature-time curves for the two preheats are shown in Figures 10 to 13.

For the plate at room temperature (Figures 10 and 11):

$$\tau_{800-500} = 5s$$

$$\tau_{1500-100} = 175s$$

For the 100°C preheated plate (Figures 12 and 13):

$$\tau_{800-500} = 8s$$

$$\tau_{1500-100} = 791s$$

4.2 Hydrogen Diffusion Analysis

Figure 14 shows a portion of the FEM mesh with some node numbers marked on it. The top and bottom surfaces (i.e. nodes 51 to 93 and nodes 4 to 10) were considered to be open surfaces, and as such, hydrogen diffusion through these surfaces was deemed to be very fast. Consequently, concentration was assumed to be zero on these two surfaces.

The iso-hydrogen contours for two preheats are shown in Figures 15 and 16. As expected, at 100°C preheat, a greater amount of hydrogen has diffused out compared to 20°C preheat.

Transient hydrogen-time plots for three nodes (viz. 45, 25 and 39 - refer to Figure 14) are shown in Figures 17 through 22 for the two preheats considered in the present study. The concentration of hydrogen is lower at these three nodes at 100°C preheat, compared to 20°C. For example, at node 45, after 100s, concentration of hydrogen at 20°C is 10ppm, compared to 5.5ppm at 100°C.

Thus, depending on the critical hydrogen concentration for cold cracking, preheat is advantageous in lowering the hydrogen concentration.

4.3 Stress Analysis

For stress analysis results, four nodes were selected (refer to Figure 14), viz. node 51 (top surface at weld centreline), node 4 (bottom surface at weld centreline), node 45 (in weld metal), and node 68 in HAZ close to the fusion boundary interface. Since the transverse stress component (σ_x) is predominantly responsible for occurrence of cracking in these tests, only this component has been plotted for two preheats and two restraint levels as functions of:

- (a) time
- (b) temperature.

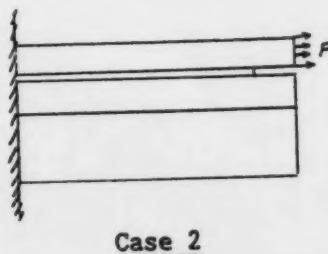
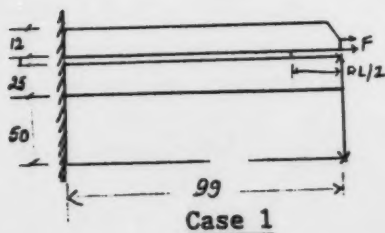
For σ_x vs. temperature plots, the cooling cycle of the weld was followed from 2000°C to room temperature. Figures 23 to 28 show the σ_x vs. temperature and time plots for restraint length of 25mm for nodes 45, 51 and 4 at 20°C preheat. Figure 29 shows the σ_x vs. time plot for node 68.

Figures 30 to 36 show the similar plots for restraint length of 100mm at room temperature (20°C). The σ_s vs. time plots for the 100°C preheat are given in Figures 37 to 50.

The general conclusion from Figures 23 through 50 is that increasing the restraint length from 25mm to 100mm decreases the maximum σ_x values. Preheat, on the other hand, has a small effect on the stress.

4.4 Restraint of the Joint

For calculation of restraint, two configurations were considered, as shown in the sketch below:



In both cases, a force F of 360N was applied on the face as shown.

For restraint length of 25mm

Average displacement in the x direction for Case 1 = .0102mm
for Case 2 = .0095mm

Hence, restraint (kg/mm^2) for Case 1 = 150 h
for Case 2 = 161 h

For restraint length of 100mm

Average displacement in the x direction for Case 1 = 0.02mm
for Case 2 = 0.187mm

Hence, restraint (kg/mm^2) for Case 1 = 76.5 h
for Case 2 = 82.0 h

The results of the stress analysis, after the weld has cooled to room temperature can be summarized as follows:

Restraint Length (mm)	Preheat (°C)	HAZ close to fusion boundary Node 68	Transverse Stress (MPa)		
			Weld Centre (Top)	Weld Centre (Bottom)	Face Side
			Node #51	Node #4	Node #45
25	20	838	720	550	800
25	100	839	700	511	790
100	20	810	575	350	730
100	100	800	560	363	702

It should be noted that the peak stresses are less than the stipulated yield strength of the materials. In the Japanese work[3], the stress levels are up to three times the yield strength in spite of the assumption of zero work hardening. It is believed the reported values are obtained as a product of the FE stress values and the elastic stress concentration factor.

5.0 RECOMMENDATIONS FOR FURTHER WORK

- 5.1 In the present study, the diffusion coefficients is assumed to be a function of temperature only. However, it is known that it is a function of plastic strain also. The effect of plastic strain on hydrogen diffusion can be obtained using the available FEM plastic strain data and the approach suggested by Kohira et al[3]. The approach involves calculating an apparent diffusivity from the following equation:

$$D_{\text{strained}} = \frac{D}{1 + 1.228 \times 10^{-3} \times V_{\text{pore}} \times \exp\left(\frac{6733}{RT}\right)}$$

and performing hydrogen diffusion analysis, using diffusivity as a function of % strain and temperature. During the stress analysis, the plastic strain data was also obtained and the hydrogen diffusion analysis can be redone to take this factor into account.

- 5.2 The analysis results have been presented in terms of stress, perpendicular to the welding direction. Thus, the calculated stresses are limited by the yield strengths of the material. Instead, a better approach might be to look at the plastic strain fields, data for which is already available as a result of the stress analysis.

- 5.3 In the present analysis, the diffusion of hydrogen is assumed to follow Fick's law. Inherent in this assumption is the fact that the medium for hydrogen diffusion is homogeneous. In the presence of plastic strain fields, the hydrogen concentration is a discontinuous function and Fick's law is not truly applicable. In order to remove this shortcoming, researchers have considered the mass diffusion equation instead. The introduction of the activity coefficient into the mass diffusion equation makes it comparatively easier to calculate the diffusion and accumulation of mass under the conditions that the property of the diffusion medium vary locally and that the stress and strains are distributed.

In this approach, the chemical potential is a continuous function, whereas concentration is discontinuous. Thus, one can model the hydrogen diffusion in a highly non-homogenous medium with abrupt changes in material properties, for example at the FZ/HAZ interface.

6.0 REFERENCES

- [1] Coe, F.R., "Welding Steels Without Hydrogen Cracking". Welding Institute, 1973.
- [2] Goldak, J.A., Chakravarti, A.P. and Bibby, M.J., "A New Finite Element Model for Welding Heat Sources". Met. Trans. B, Volume 15B, June 1984, pp.299-305.
- [3] Kohira, K., Yatake, T. and Yurioka, N., "A Numerical Analysis of the Diffusion and Trapping of Hydrogen in Steels and Steel Weldments". JWS, Vol. 43, No. 9, 1974.

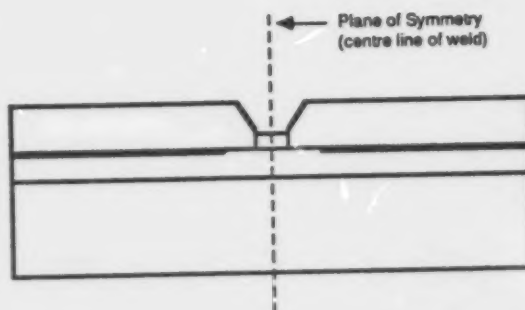
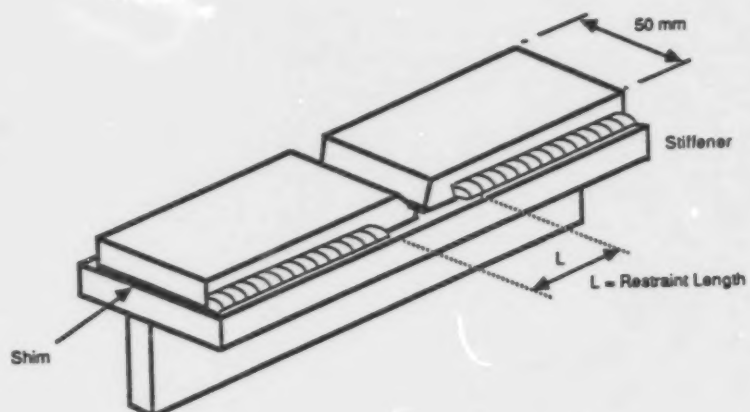


Figure 1 WIC Restraint Test

HYDROGEN DIFFUSIVITY

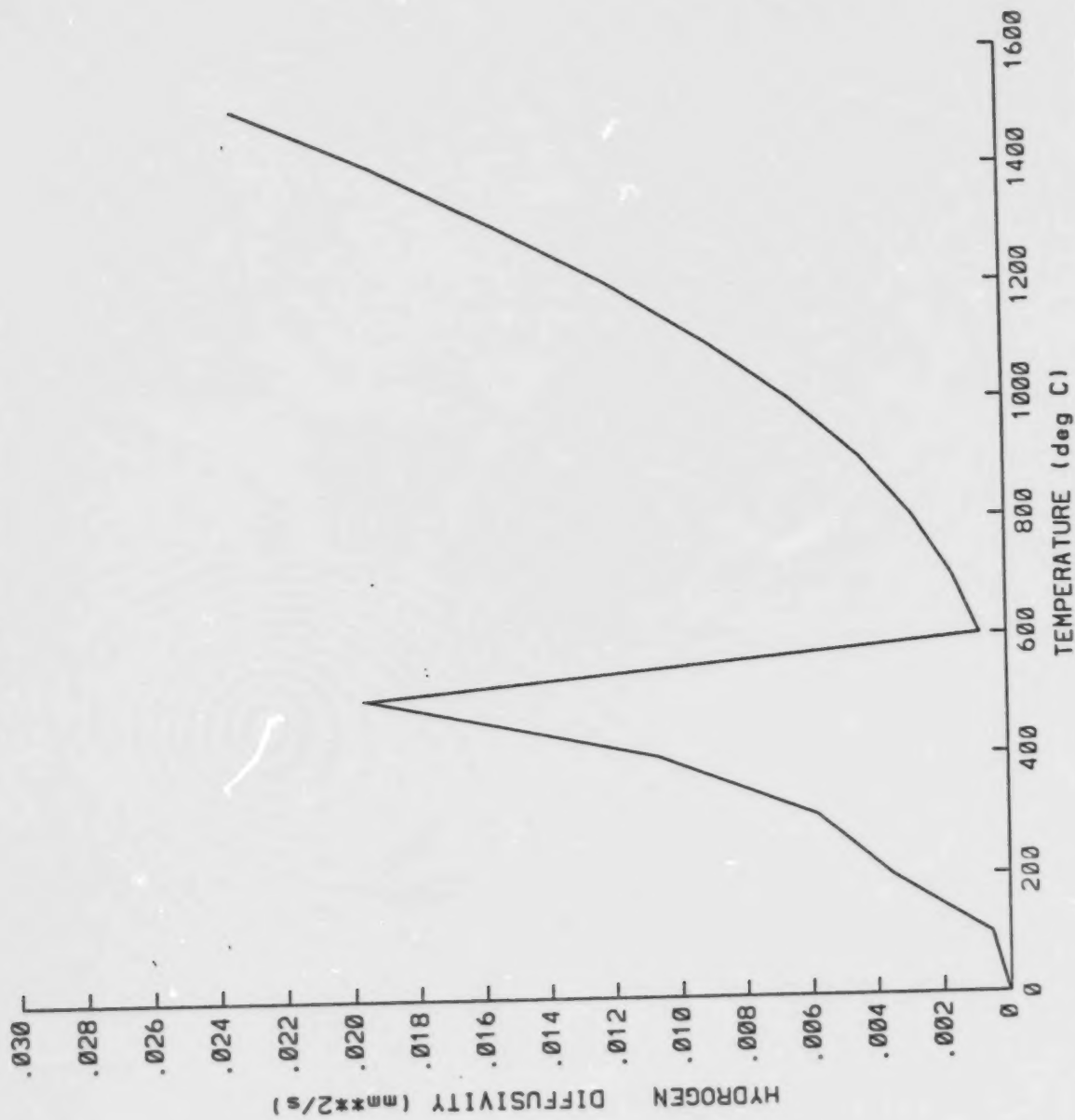


FIGURE 2: HYDROGEN DIFFUSIVITY AS A FUNCTION OF TEMPERATURE

YIELD STRENGTH - TEMPERATURE PLOT

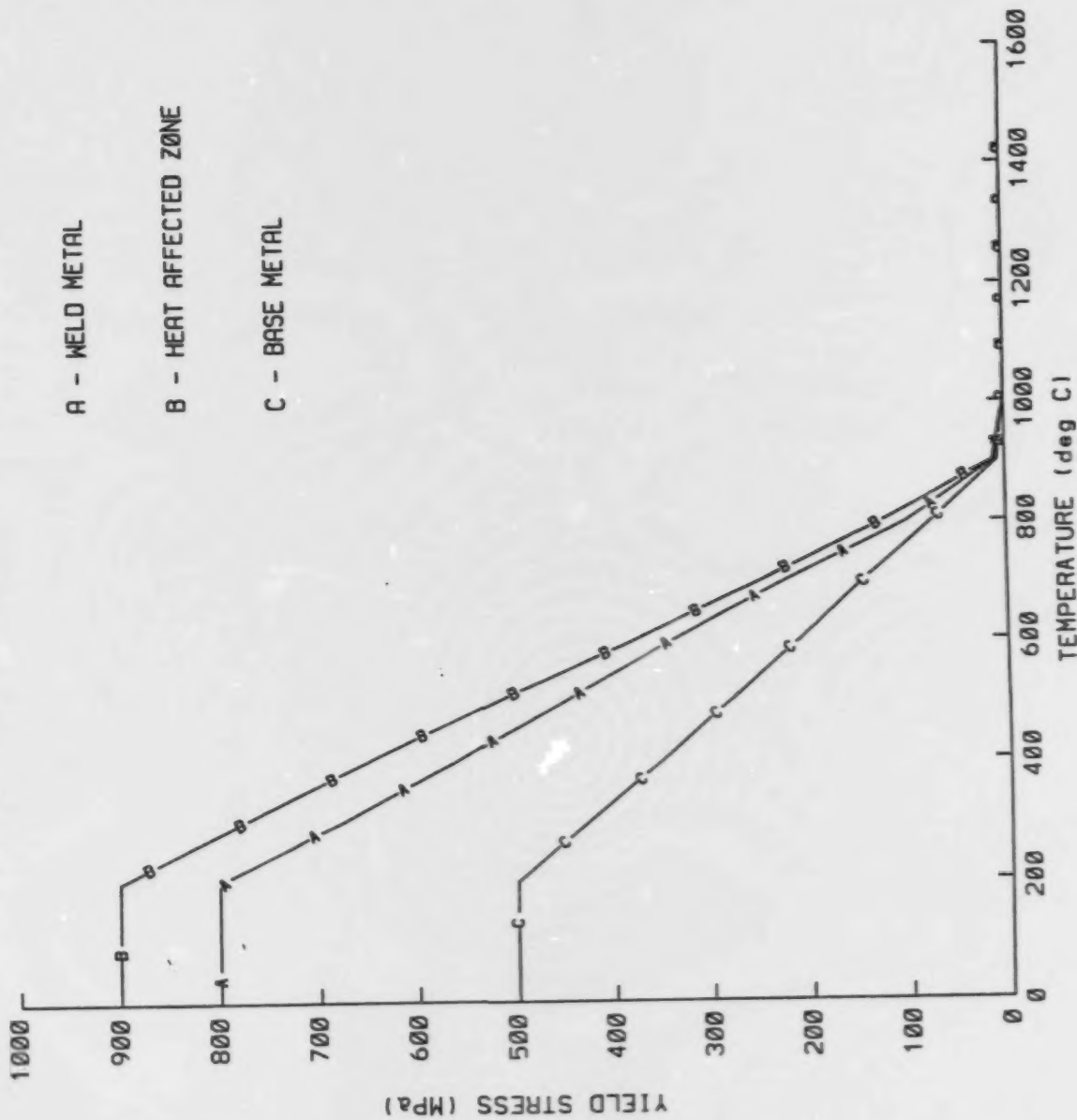


FIGURE 3: YIELD STRENGTHS OF BASE METAL, WELD METAL AND HEAT AFFECTED ZONES.

THERMAL CONDUCTIVITY OF PLAIN CARBON ST

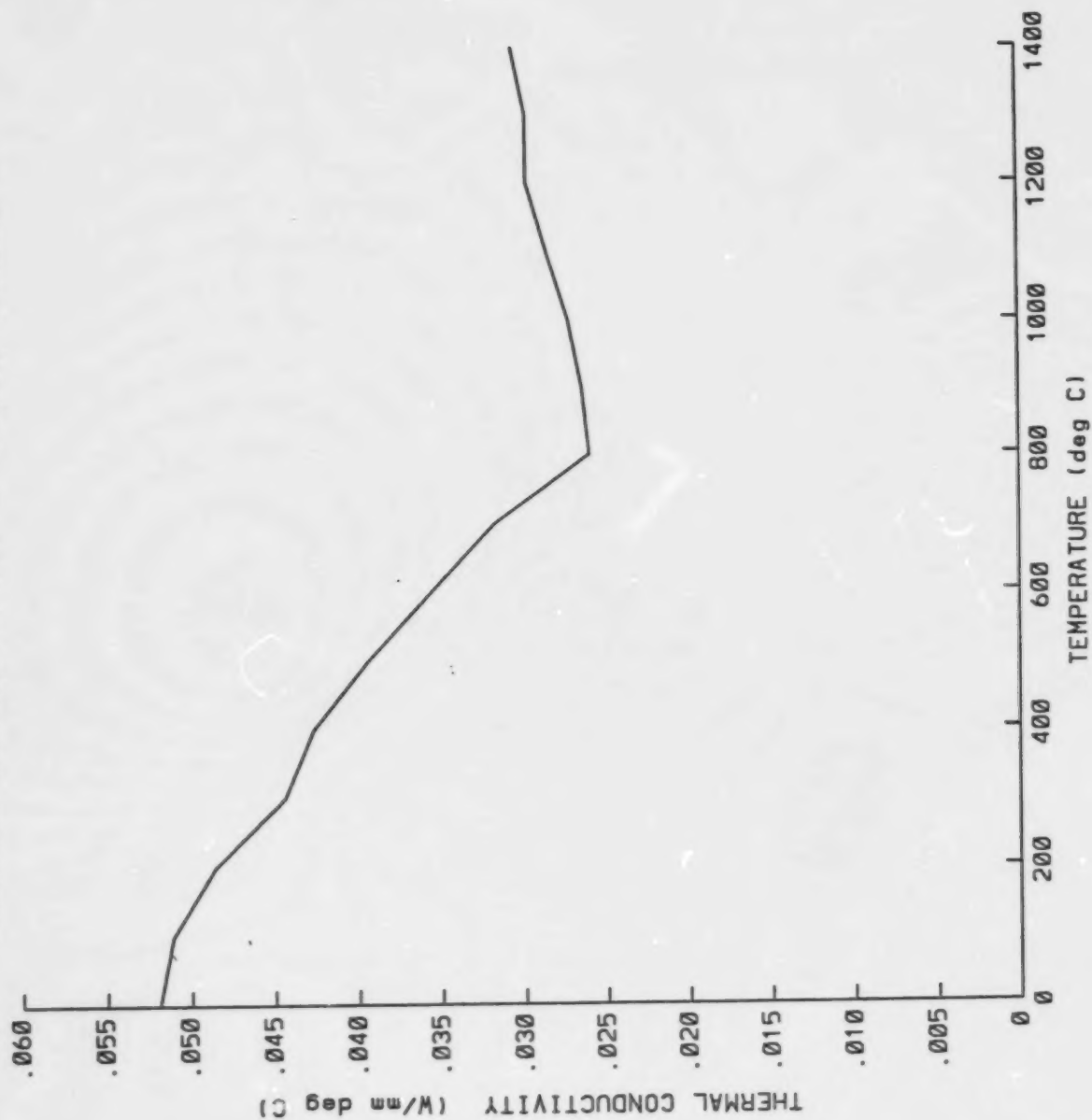


FIGURE 4: THERMAL CONDUCTIVITY VERSUS TEMPERATURE PLOT.

VOLUMETRIC SP. HEAT OF PLAIN CARBON STEEL

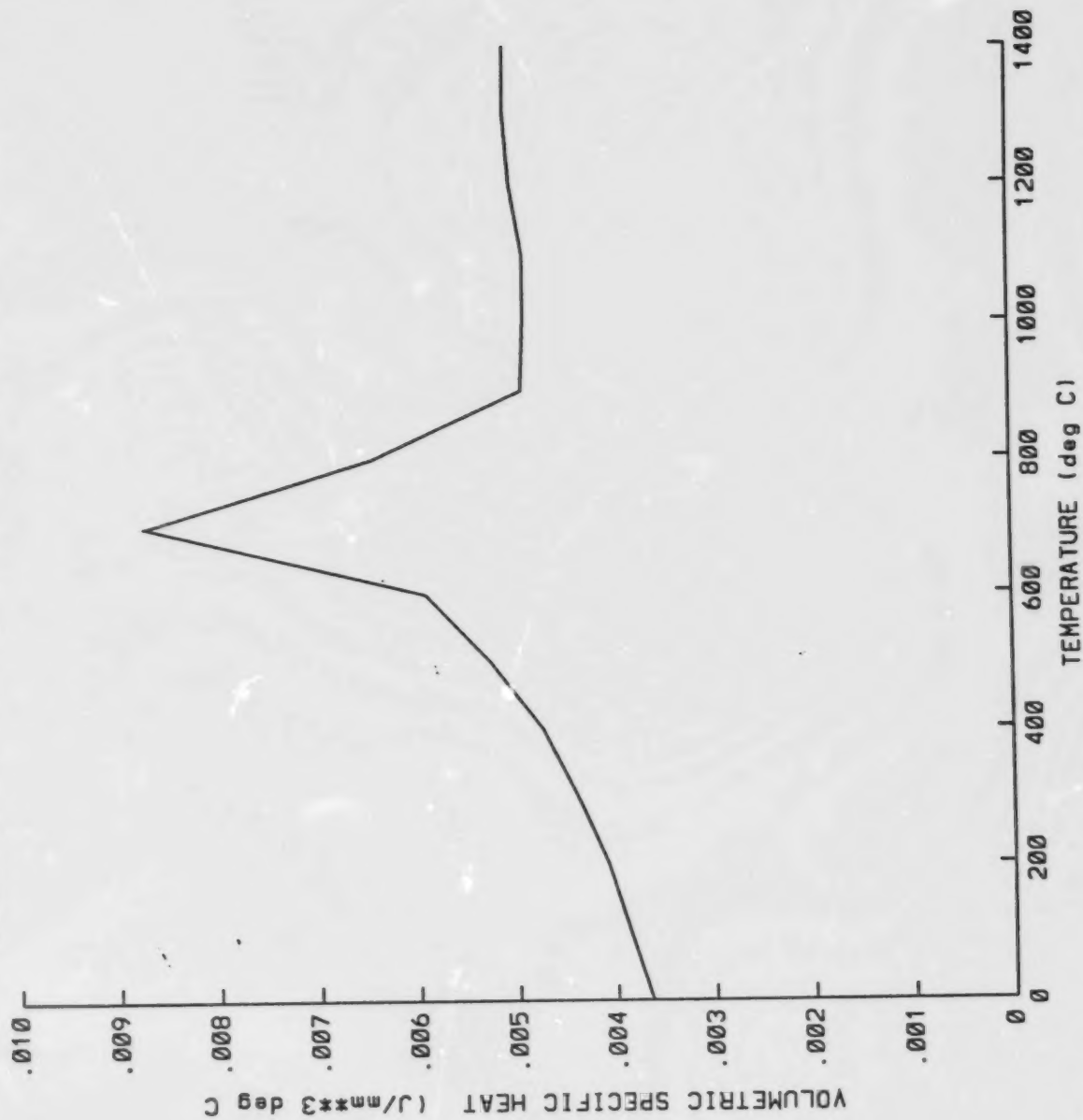


FIGURE 5: TEMPERATURE DEPENDENCE OF VOLUMETRIC SPECIFIC HEAT.

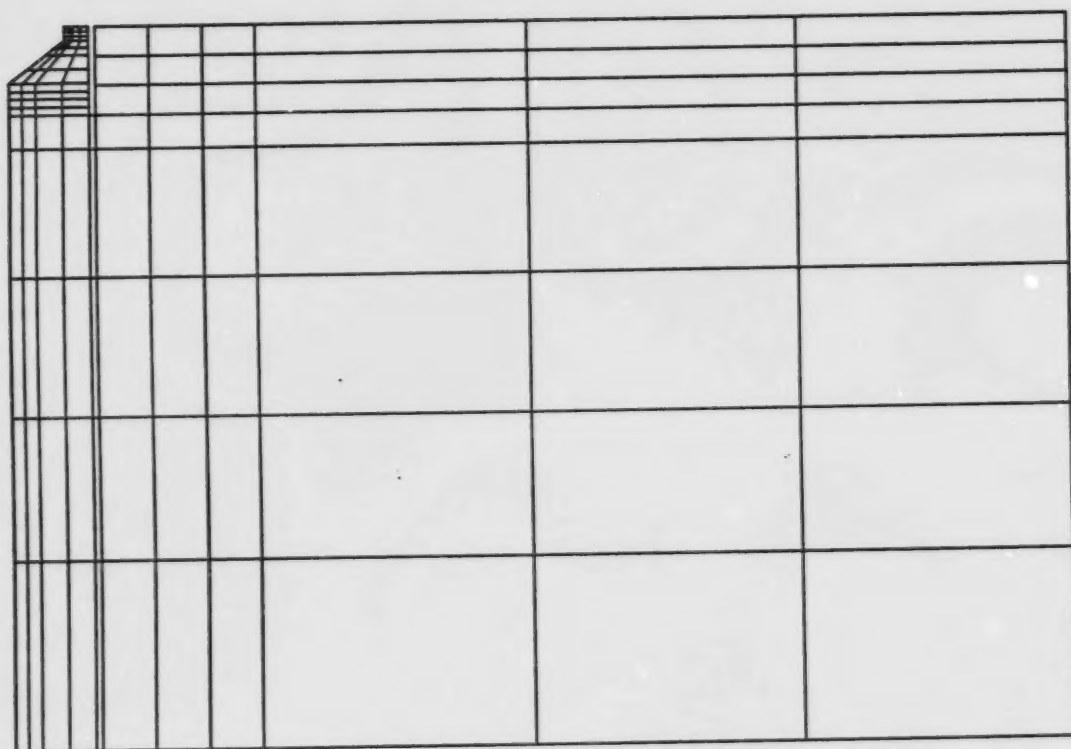


FIGURE 6: FINITE ELEMENT MESH REPRESENTATION OF WTC TEST SETUP

Heat Transfer Co-efficient = $2.4 \times 10^{-5} \text{ w/mm}^2$

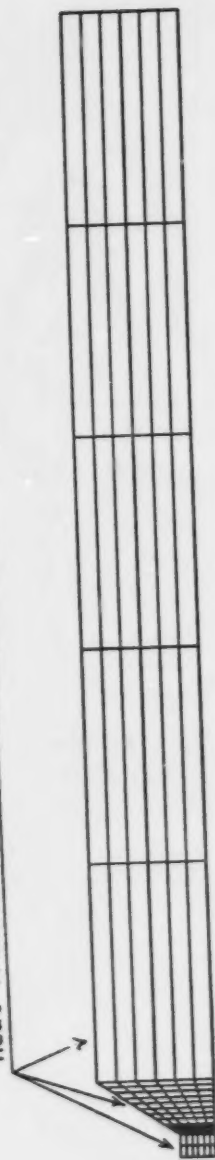


FIGURE 7: A PART OF THE FINITE ELEMENT MESH

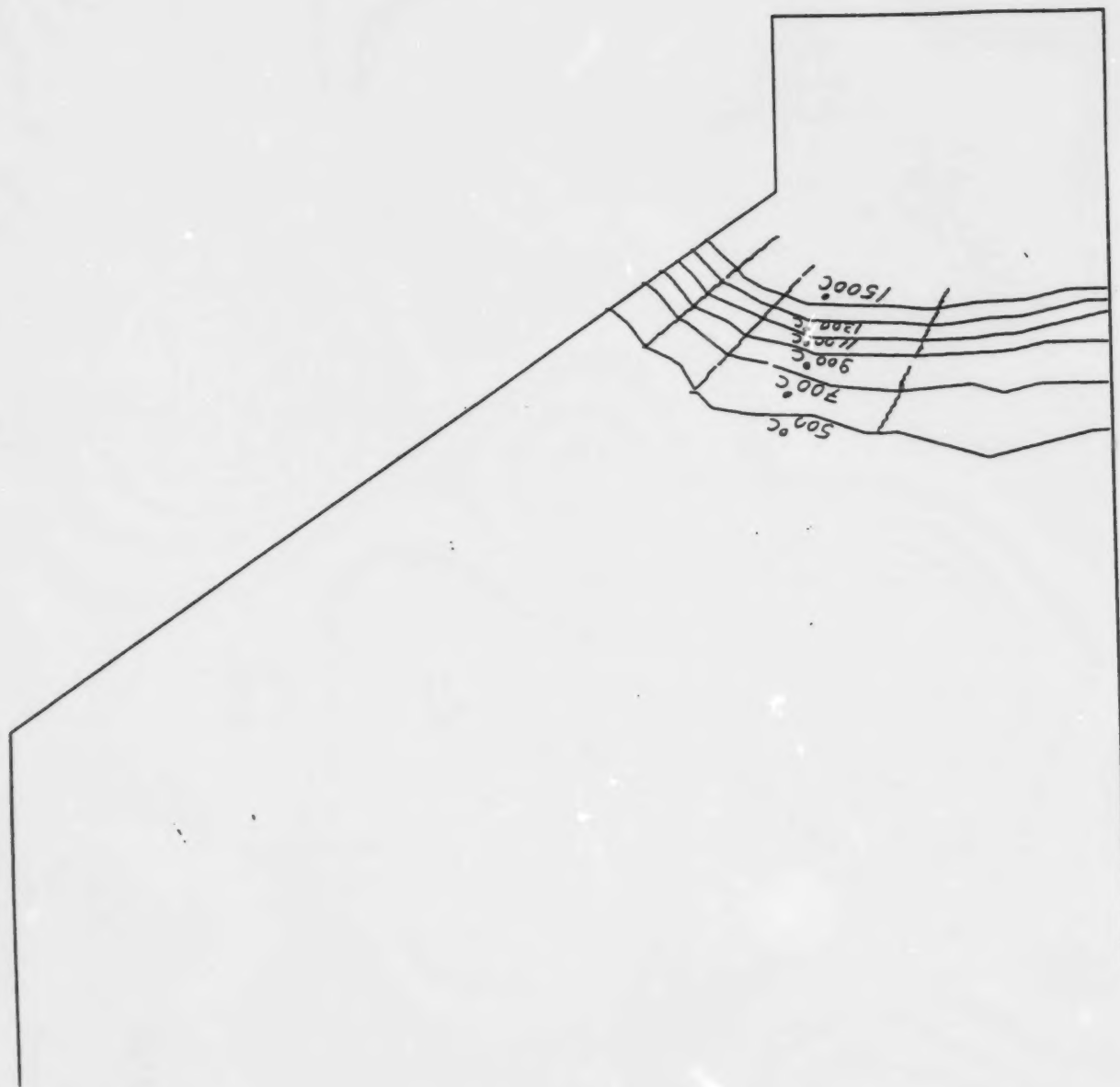


FIGURE 8: 1500-500 DEG C CØNTØUR PLØTS. PLATE AT RØØM TEMPERATURE.

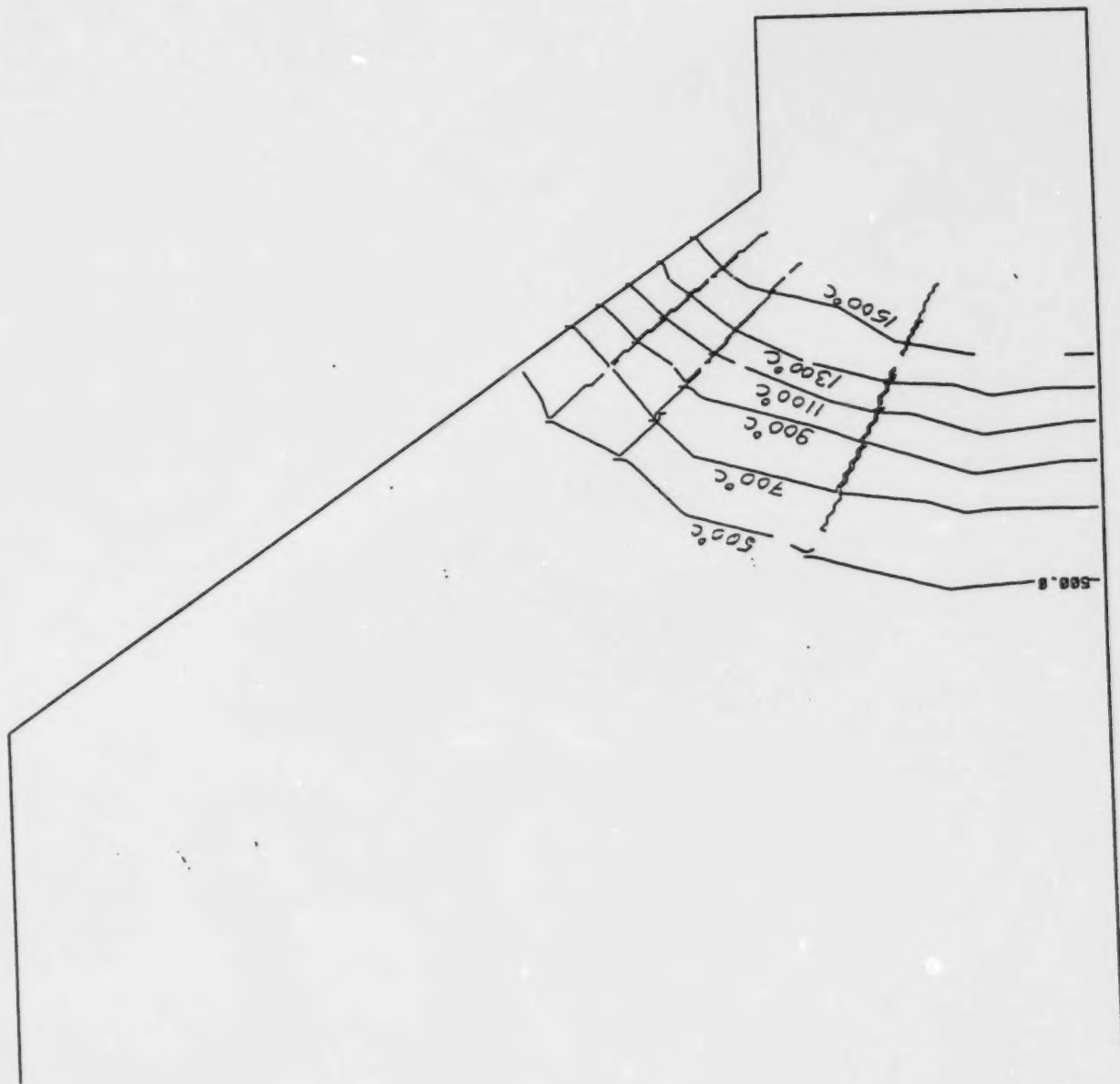


FIGURE 9: 1500-500 DEG C CØNTØUR PLØTS. PLATE AT 100 DEG C PREHEAT.

TEMPERATURE - TIME PLOT

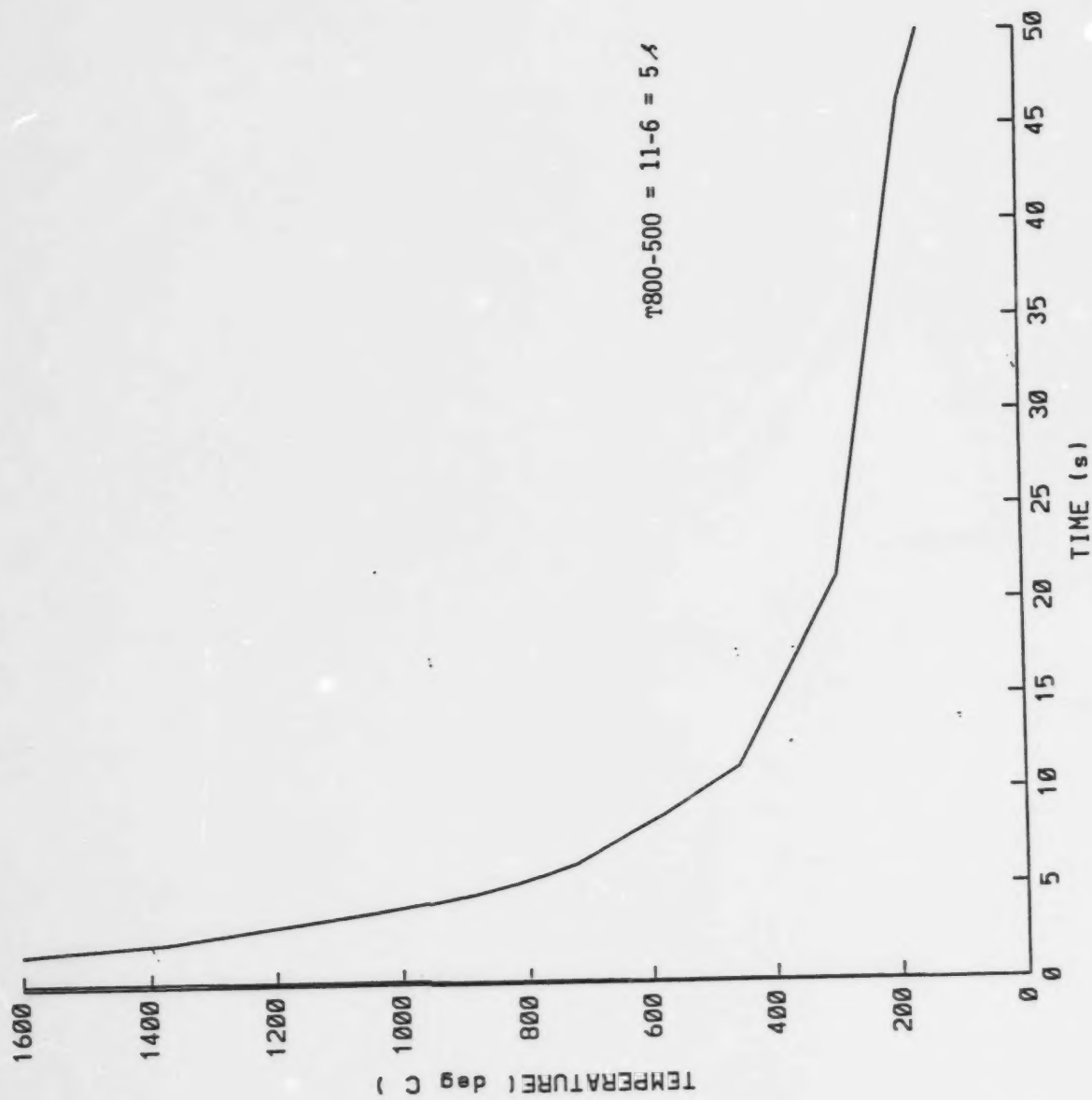


FIGURE 10: TEMPERATURE AT THE WELD CENTRE (ROOM TEMPERATURE).

TEMPERATURE - TIME PLOT

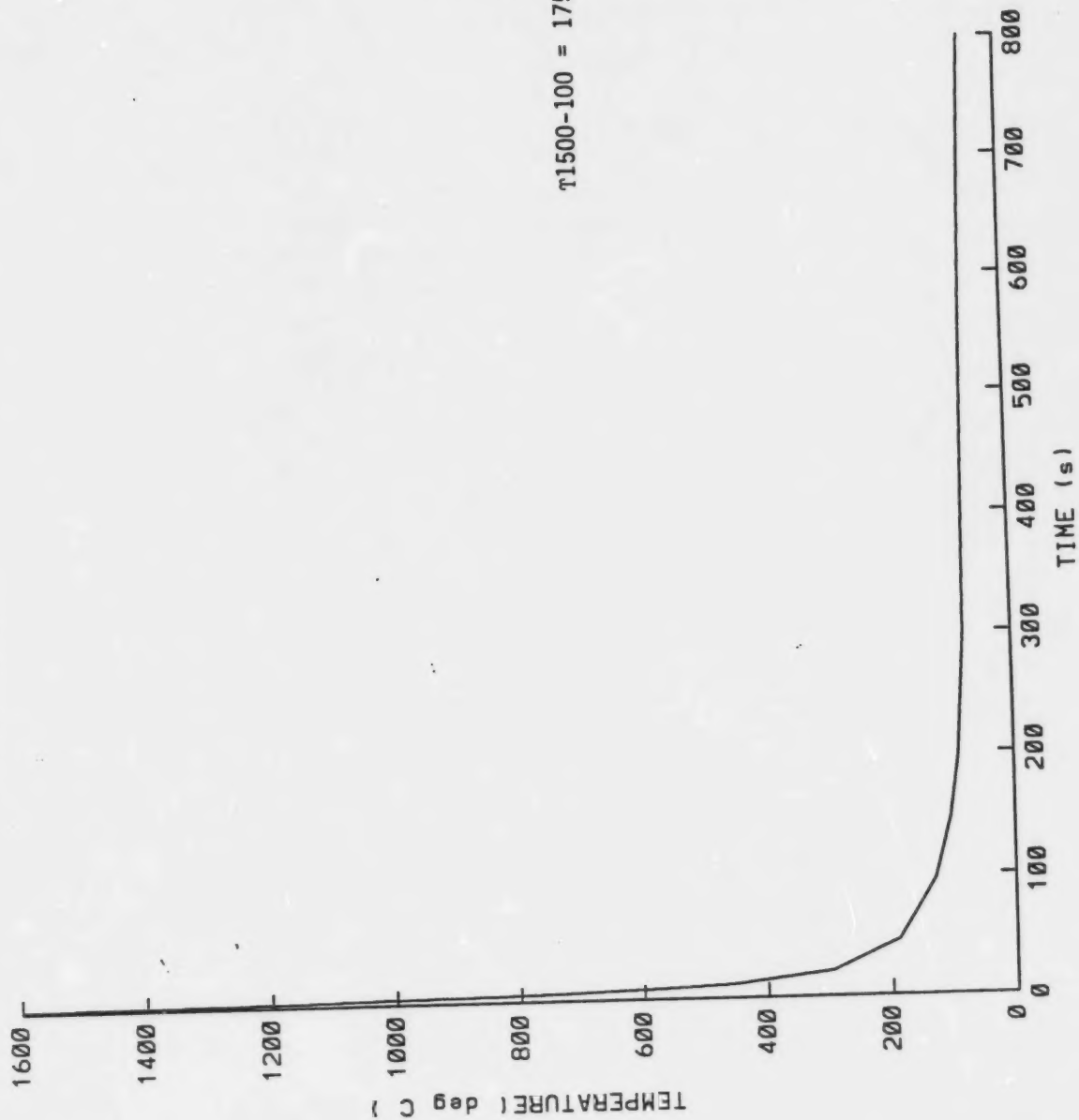


FIGURE 11: TEMPERATURE AT THE WELD CENTRE (ROOM TEMPERATURE).

TEMPERATURE - TIME PLOT

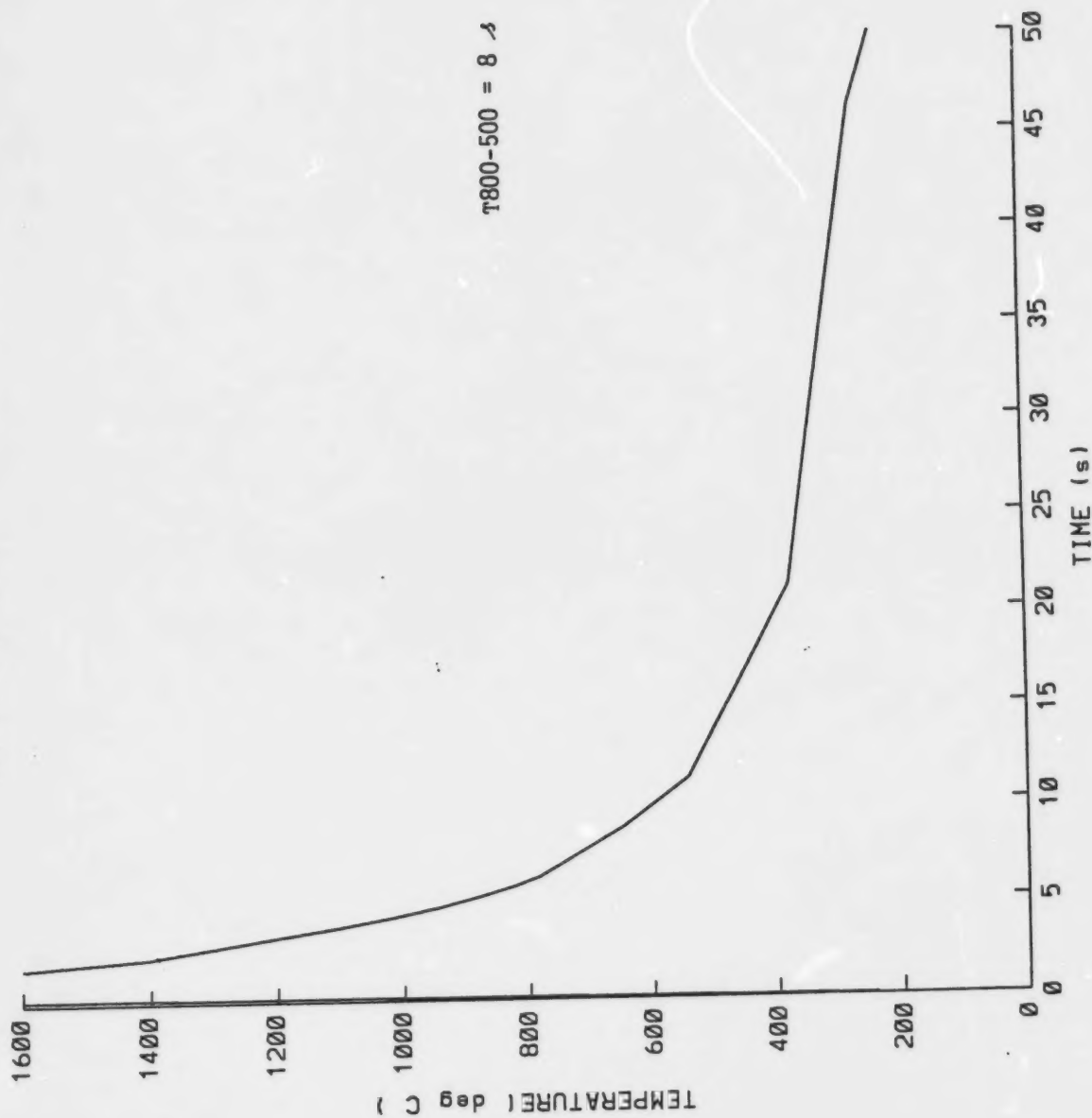


FIGURE 12: TEMPERATURE AT THE WELD CENTRE (100 PREHEAT).

TEMPERATURE - TIME PLOT

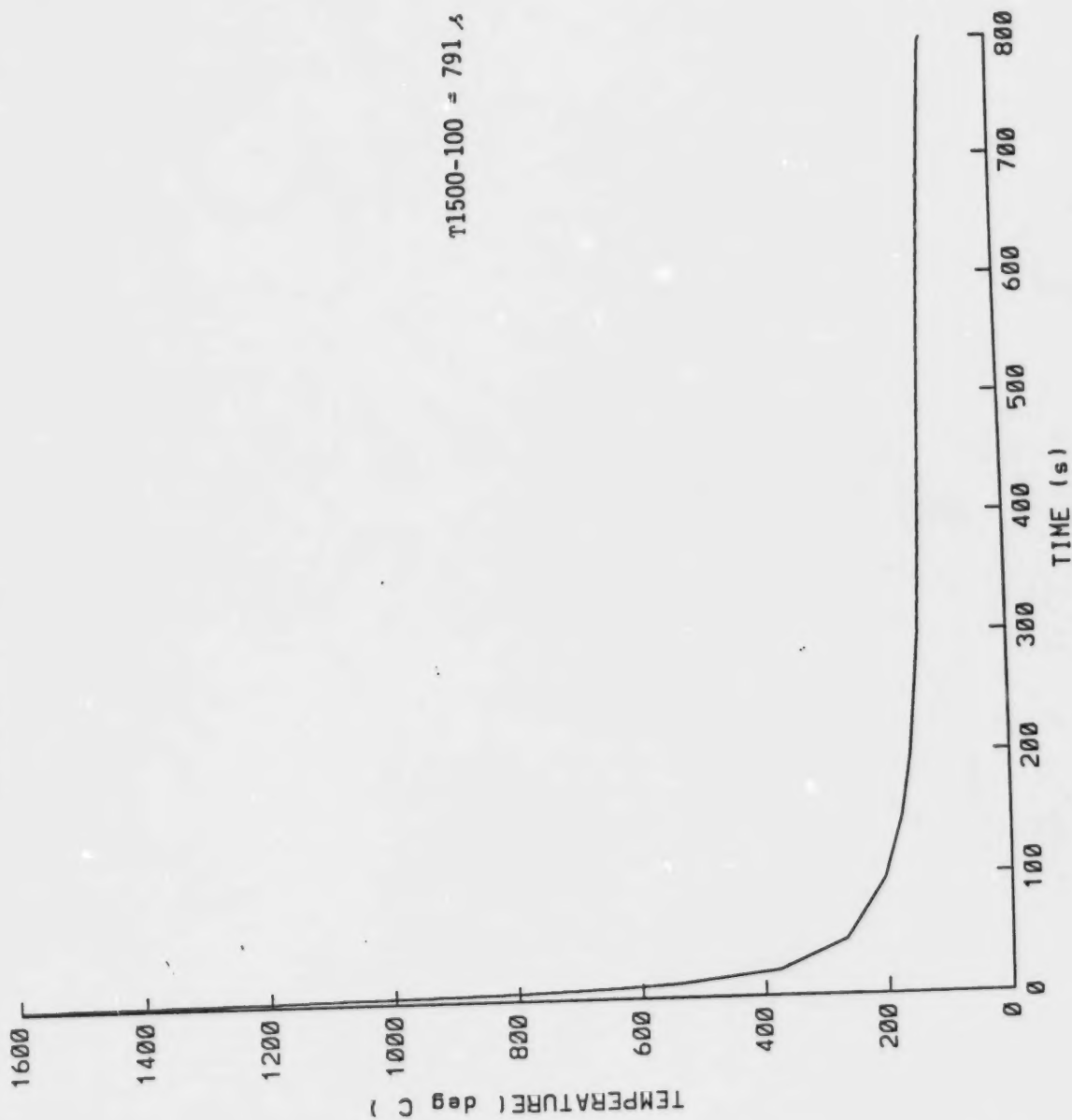


FIGURE 13: TEMPERATURE AT THE WELD CENTRE (100 PREHEAT).

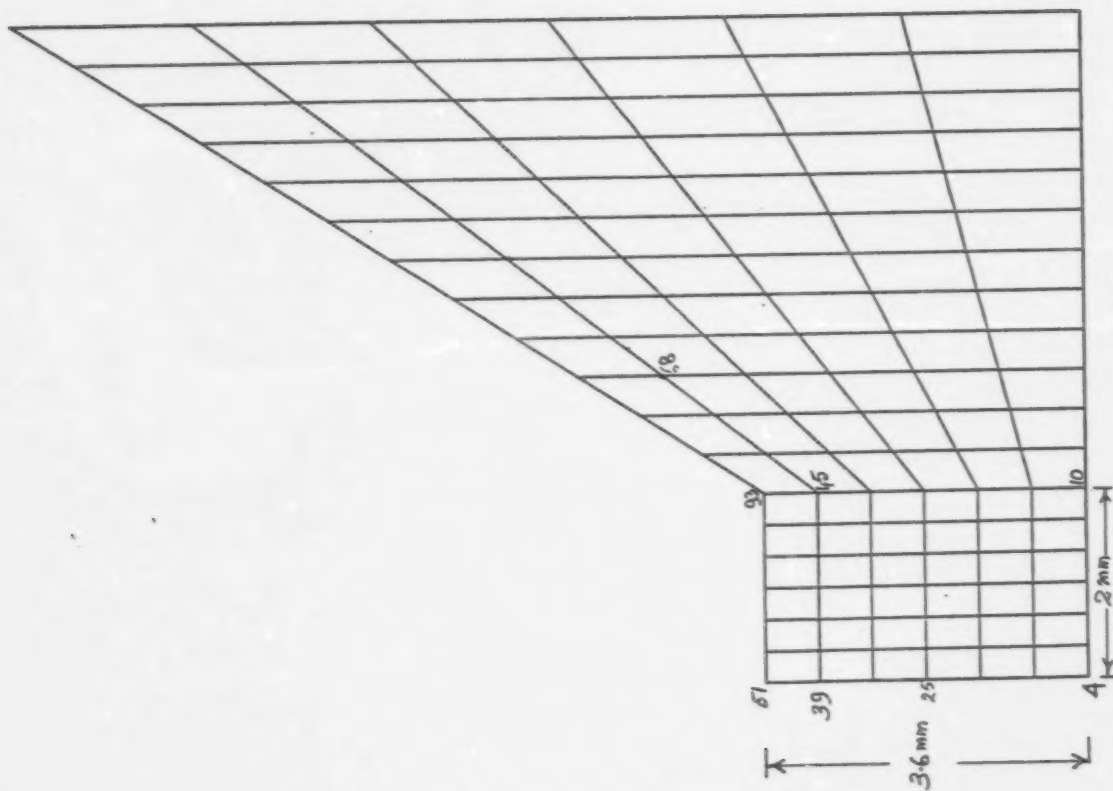


FIGURE 14: LOCATION OF SOME NODE NUMBERS

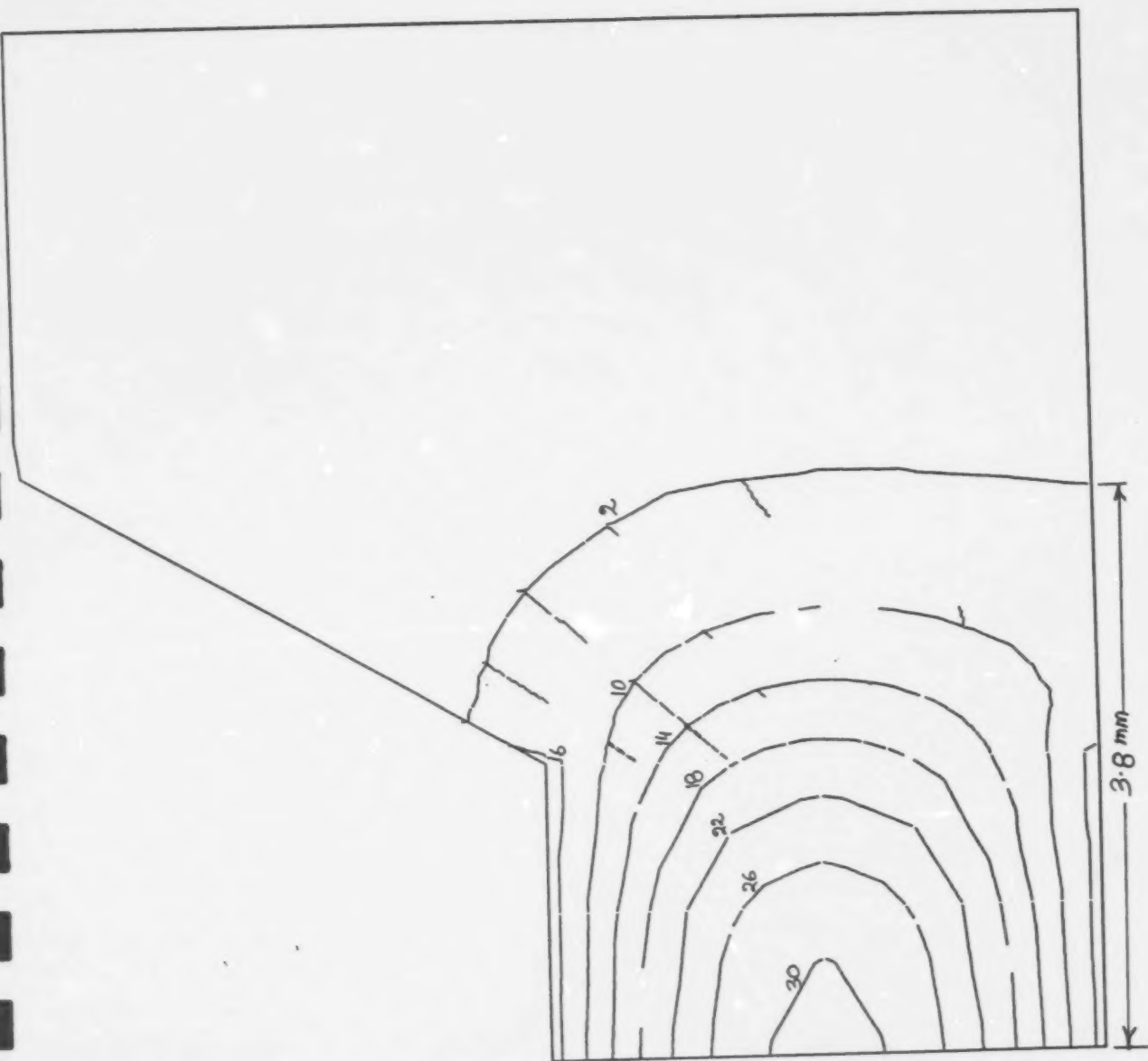


FIGURE 15: HYDROGEN CØNTØUR PLØTS (RØØM TEMP).

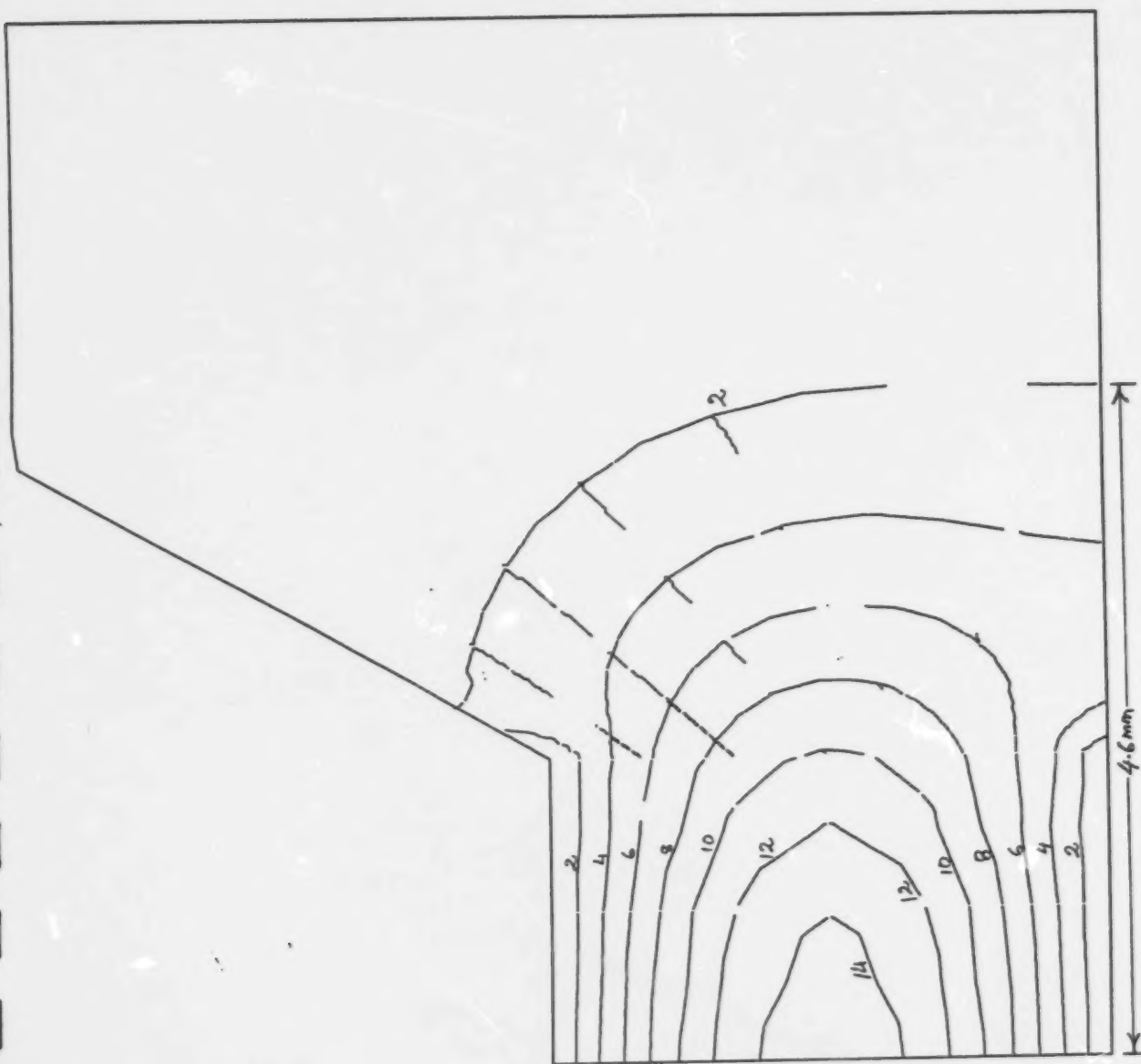


FIGURE 16: HYDRØGEN CØTØURS (100 DEG C PREHEAT).

HYDROGEN - TIME PLOT

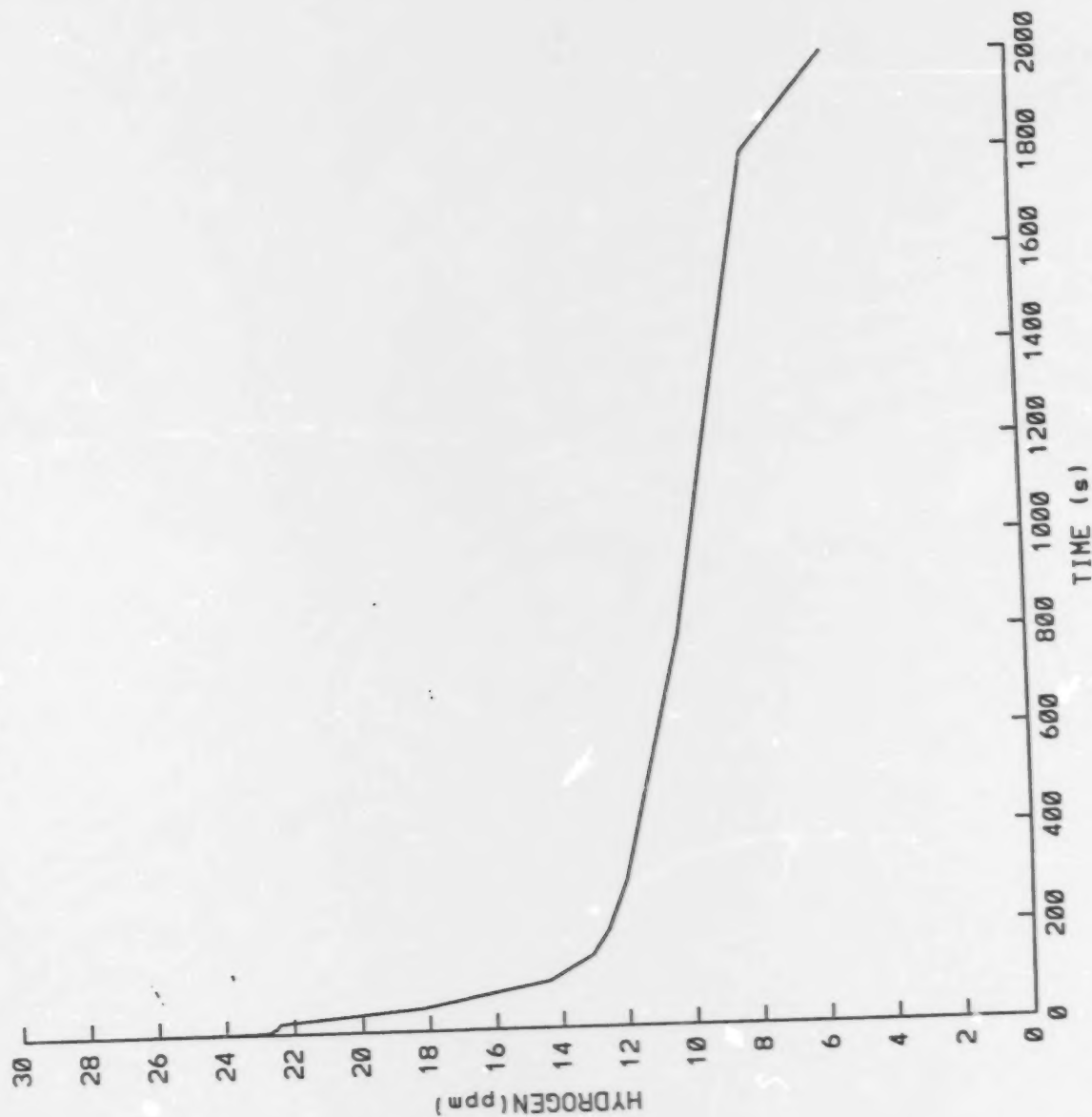


FIGURE 17: TOP-BOT H2 DIST. AT NODE 45 (ROOM TEMP).

HYDROGEN - TIME PLOT

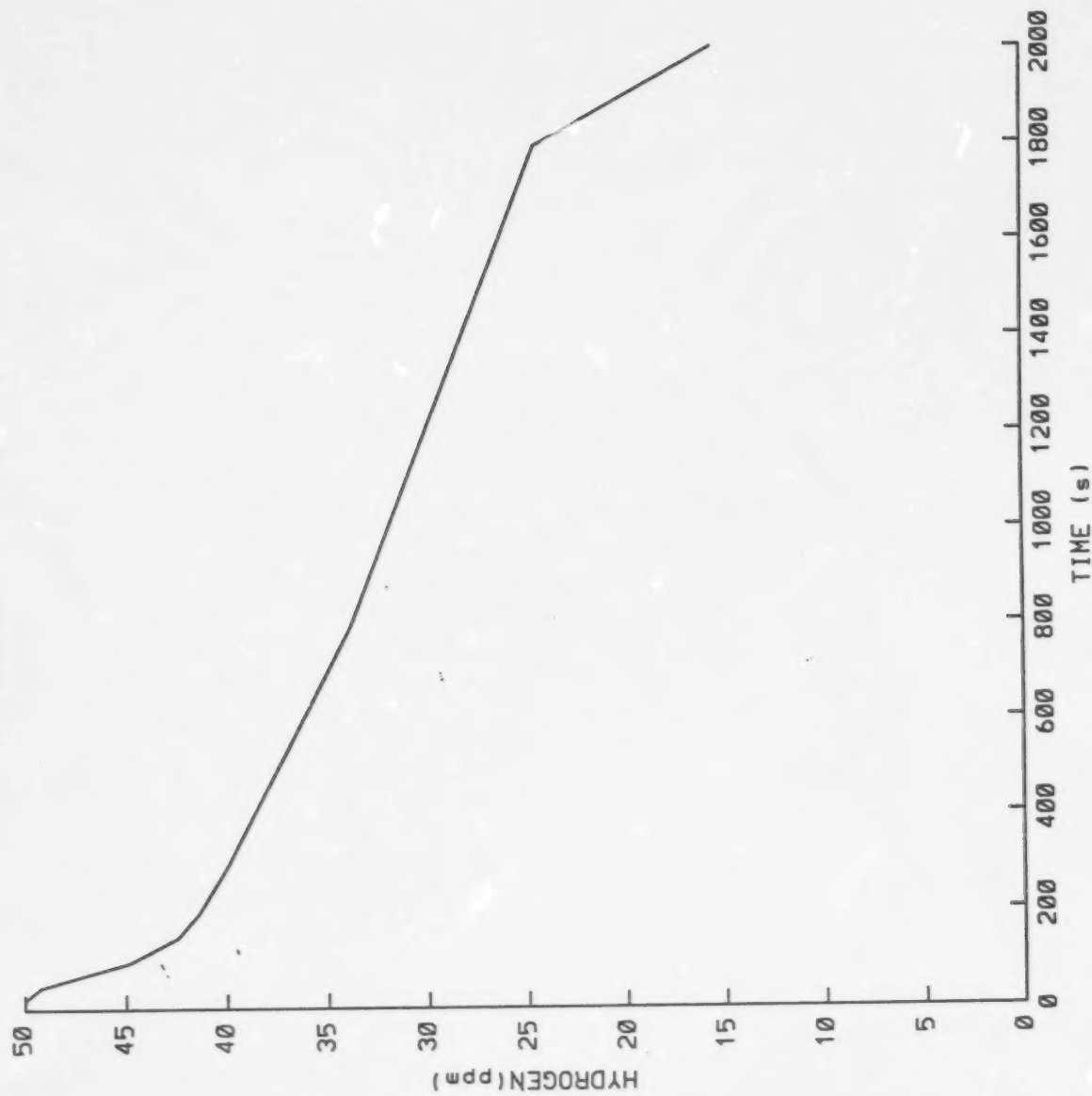


FIGURE 18: H2 DIST AT NØDE 25(ROOM TEMP), H2 LØSS FRØM TØP, BØT.

HYDROGEN - TIME PLOT

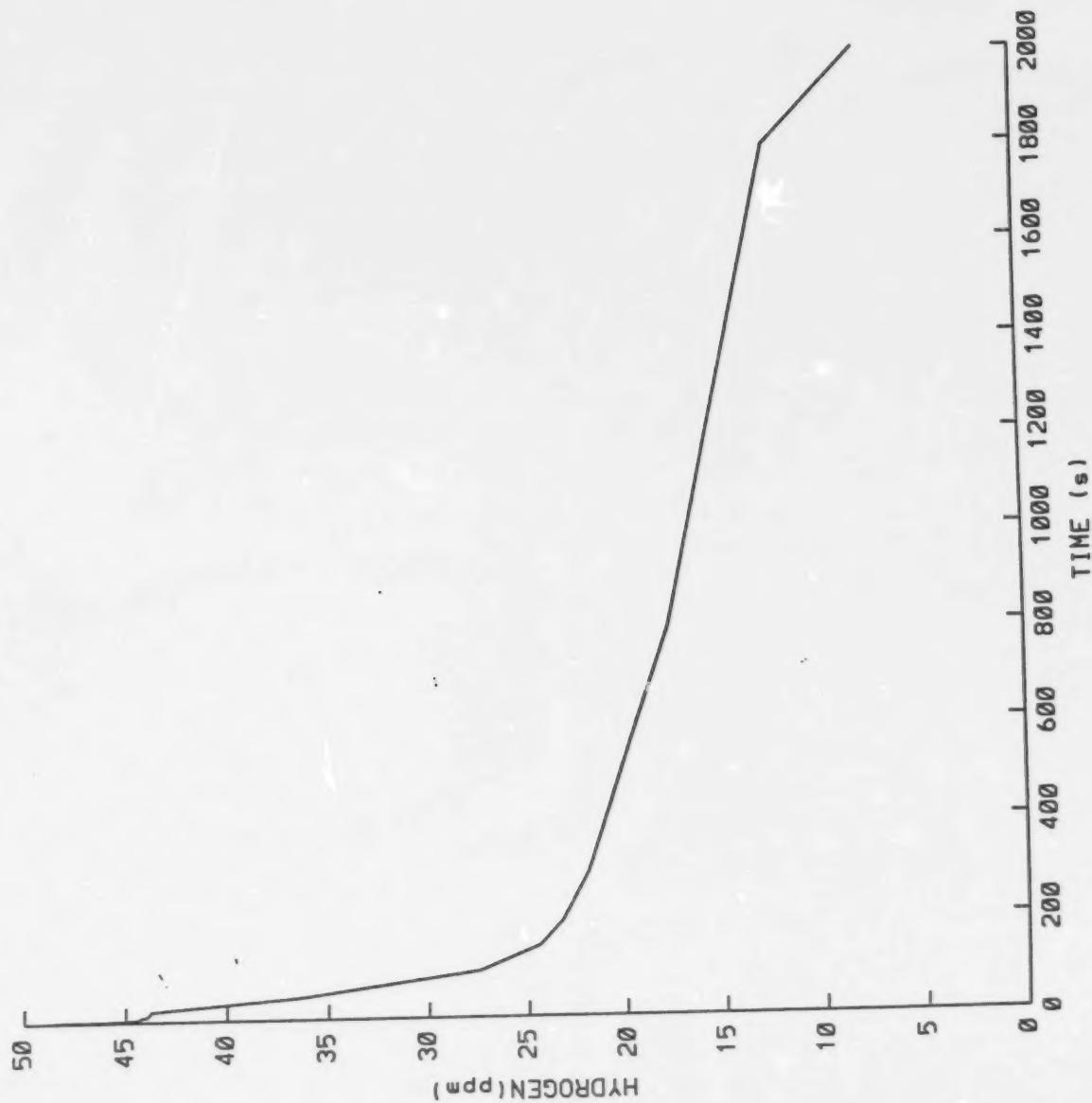


FIGURE 19: HYDROGEN DISTRIBUTION AT NODE 39 AT ROOM TEMPERATURE.

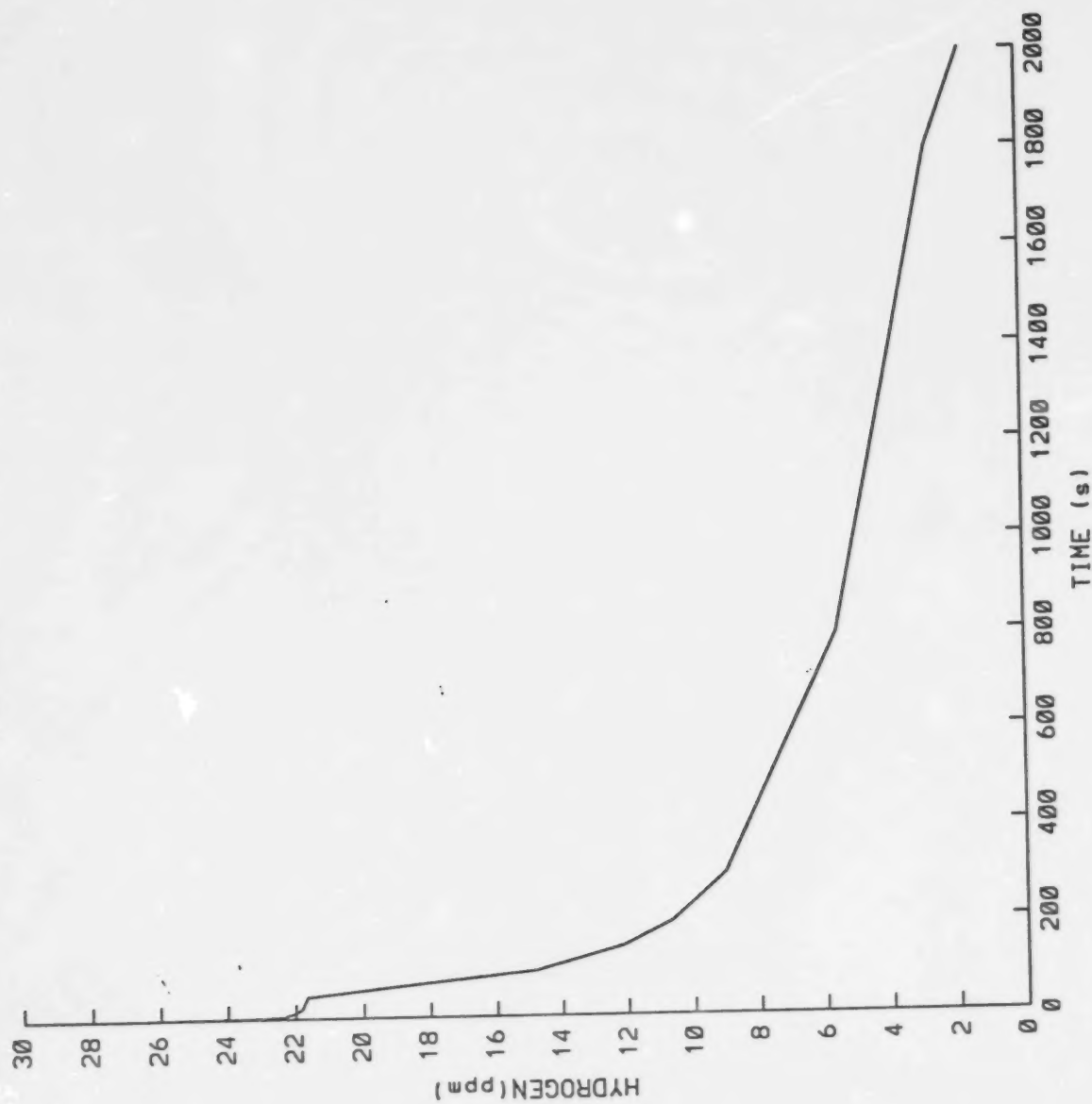


FIGURE 20: TOP BOT H2 AT NODE 45 (100 DEG C PREHEAT)

HYDROGEN - TIME PLOT

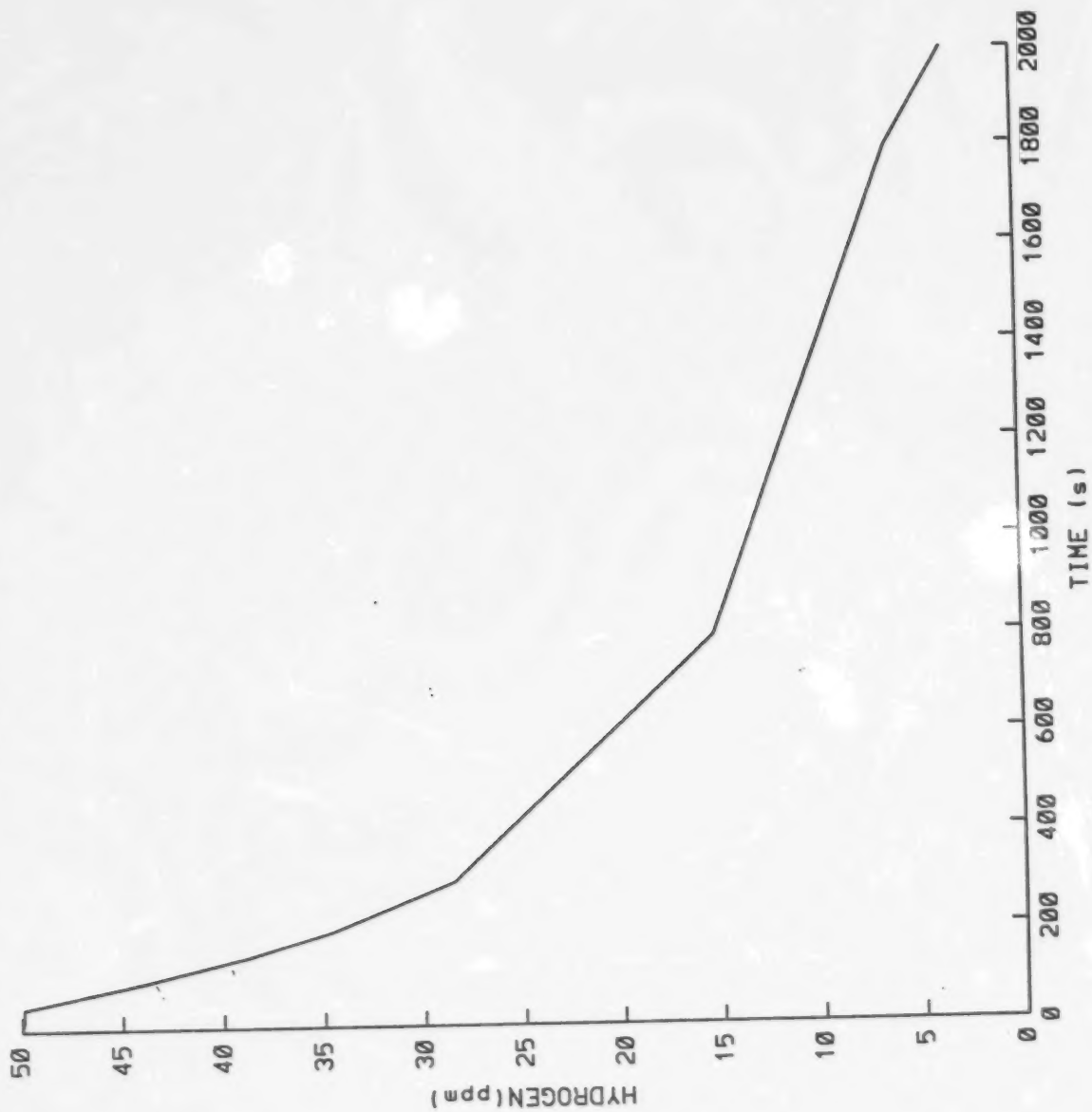


FIGURE 21: HYDROGEN-TIME PLOT (100 DEG .TOP BOT AT 0 PPM).NODE 25.

HYDROGEN - TIME PLOT

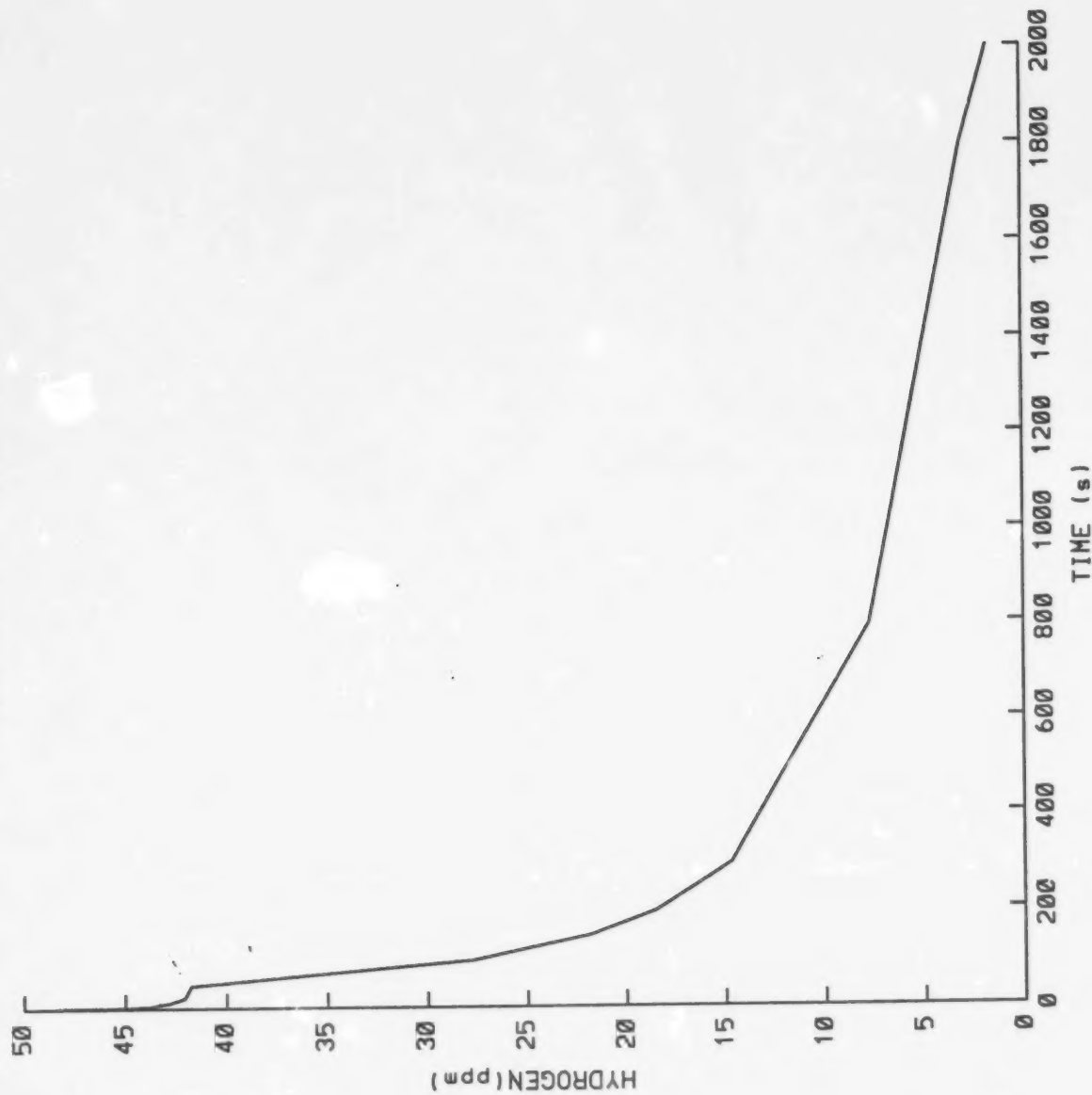


FIGURE 22: HYDROGEN DISTRIBUTION AT NODE 39 (PREHEAT=100 DEG C).

SIGMAX - TEMP. PLOT

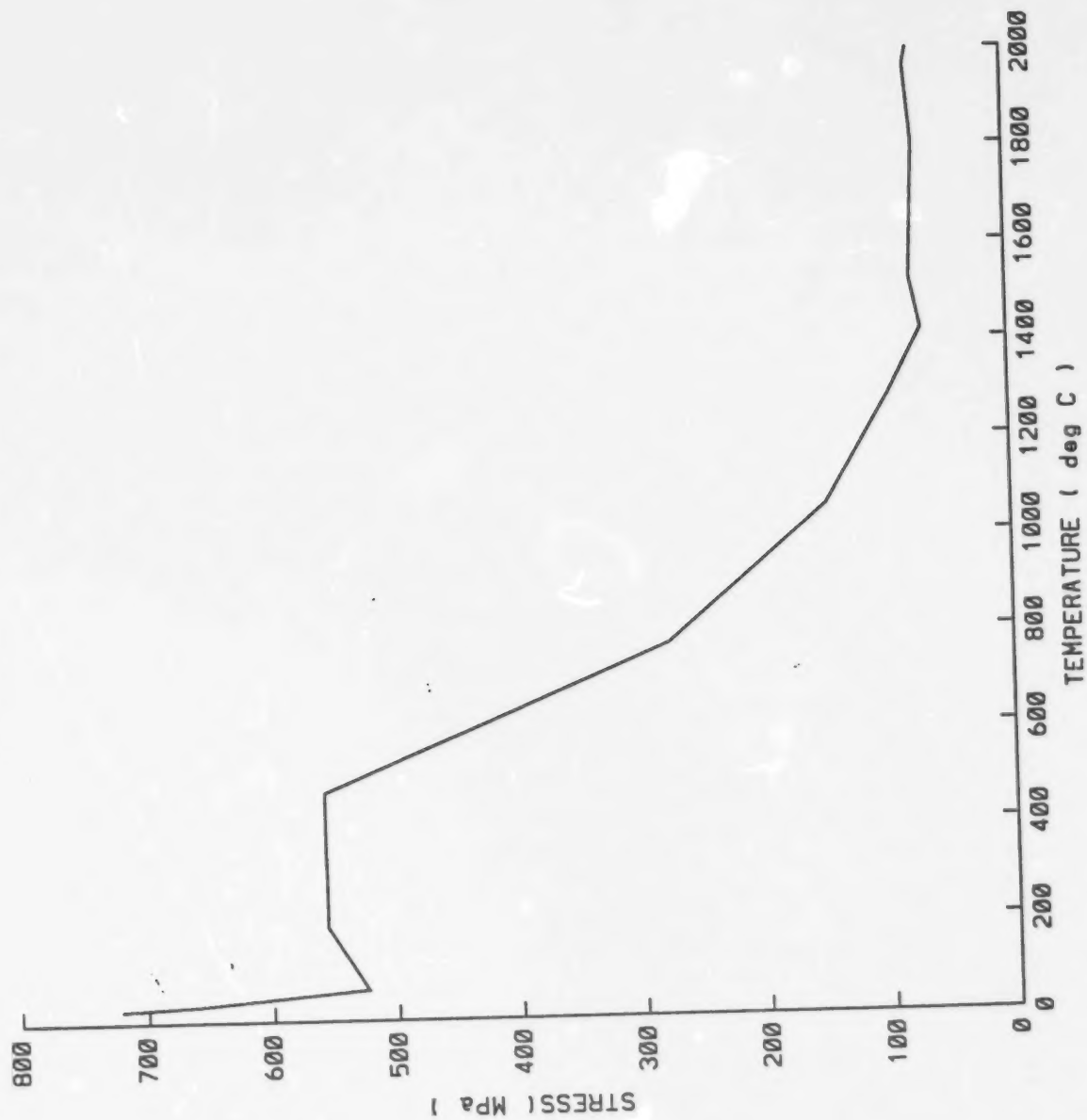


FIGURE 23: TRANSVERSE STRESS AT WELD CENTRE (TOP) AT ROOM TEMP (RL=25 MM)

SIGMAX - TEMP. PLOT

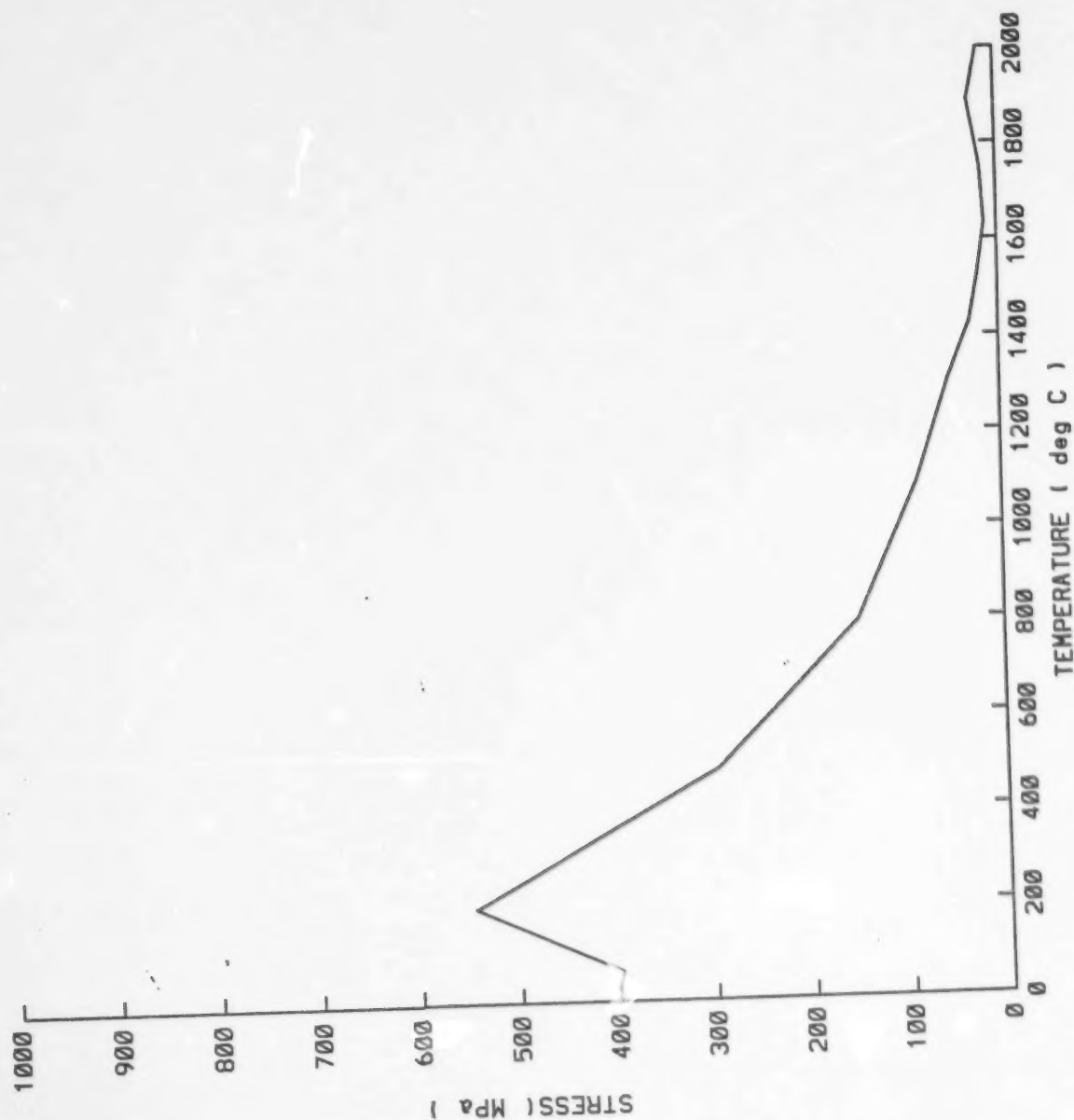


FIGURE 24: TRANSVERSE STRESS AT WELD CENTRE (BØTTØM), ROOM TEMP (RL=25 MM)

SIGMAX - TEMP. PLOT

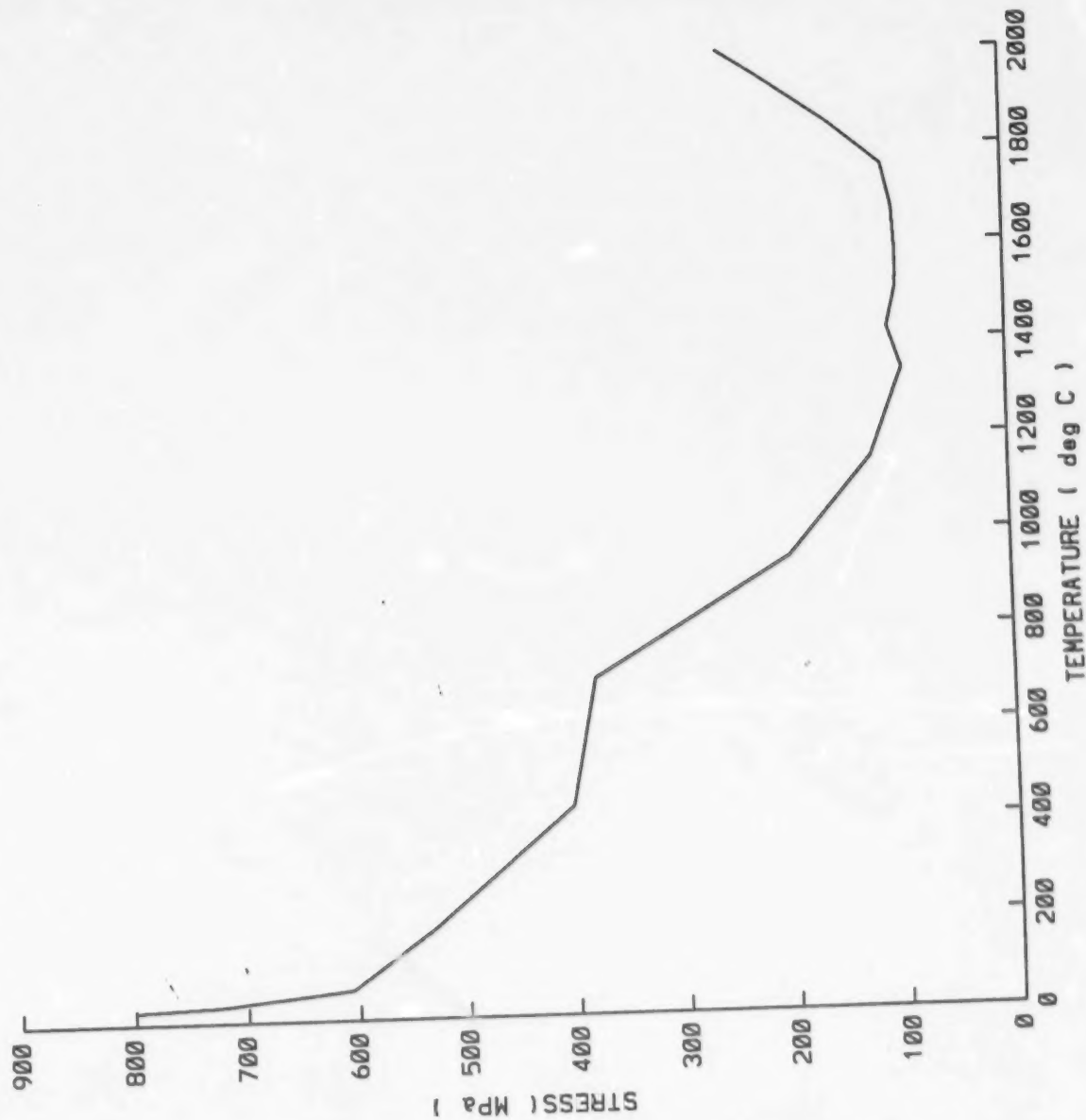


FIGURE 25: TRANSVERSE STRESS(NODE 45), ROOM TEMP, RL = 25 MM.

SIGMAX - TIME PLOT

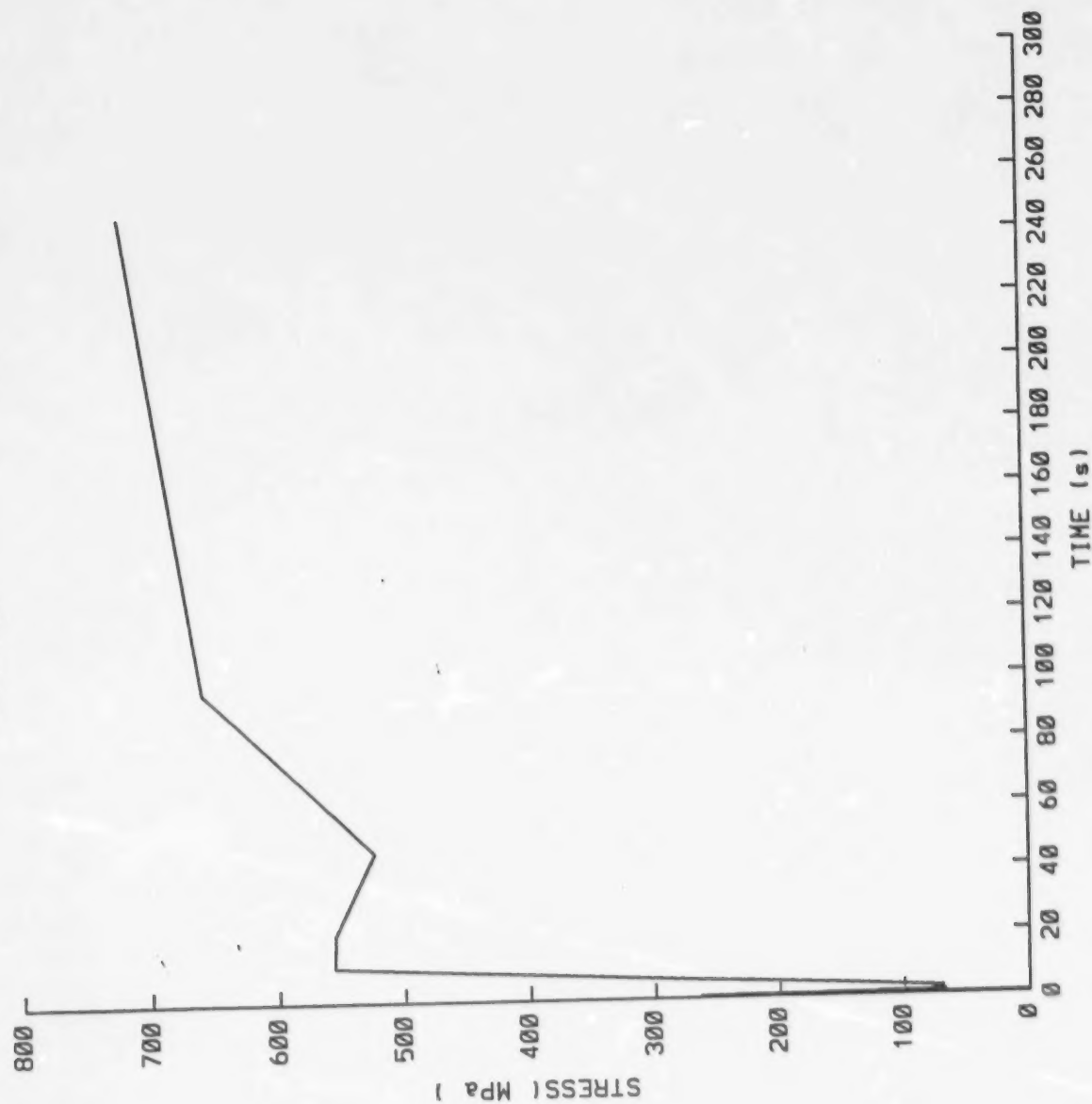


FIGURE 26: TRANSVERSE STRESS AT WELD CENTRE(TOP) AT ROOM TEMP(RL=25 MM)

SIGMAX - TIME PLOT

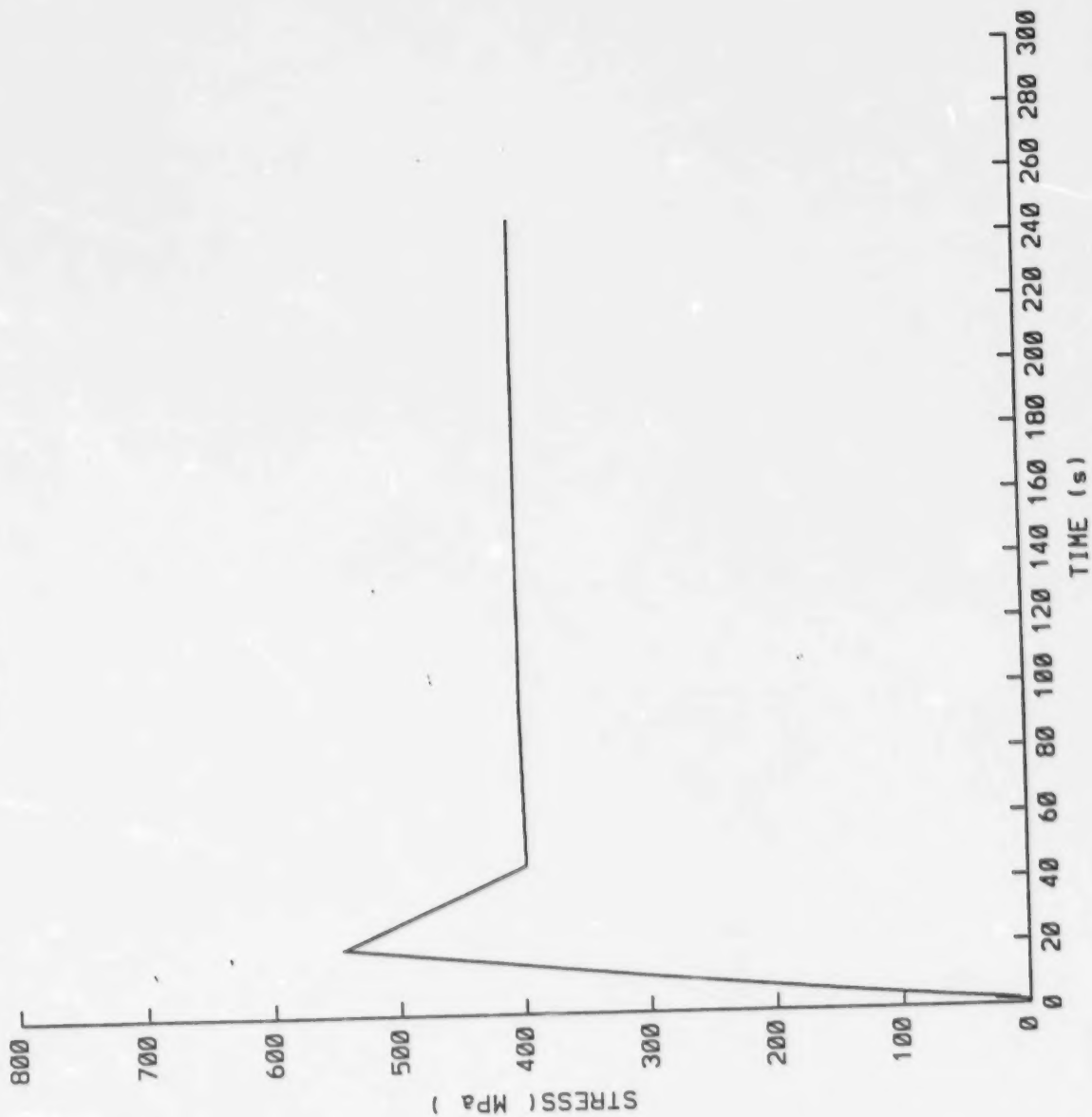


FIGURE 27: TRANSVERSE STRESS AT WELD CENTRE (BOTTOM). ROOM TEMP (RL=25).

SIGMAX - TIME PLOT

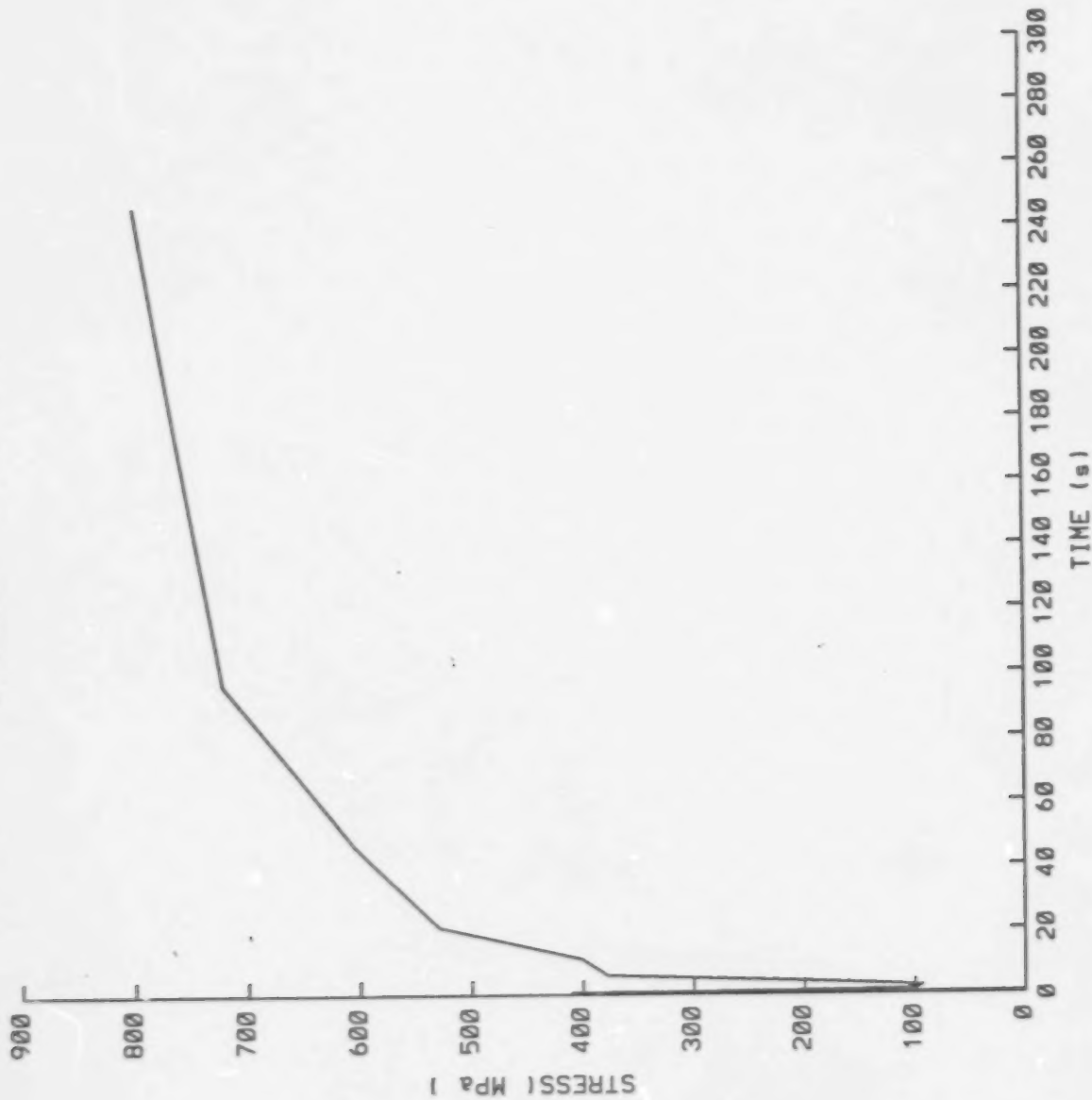


FIGURE 28: TRANSVERSE STRESS AT NODE 45, ROOM TEMP., RL = 25 MM.

SIGMAX - TIME PLOT

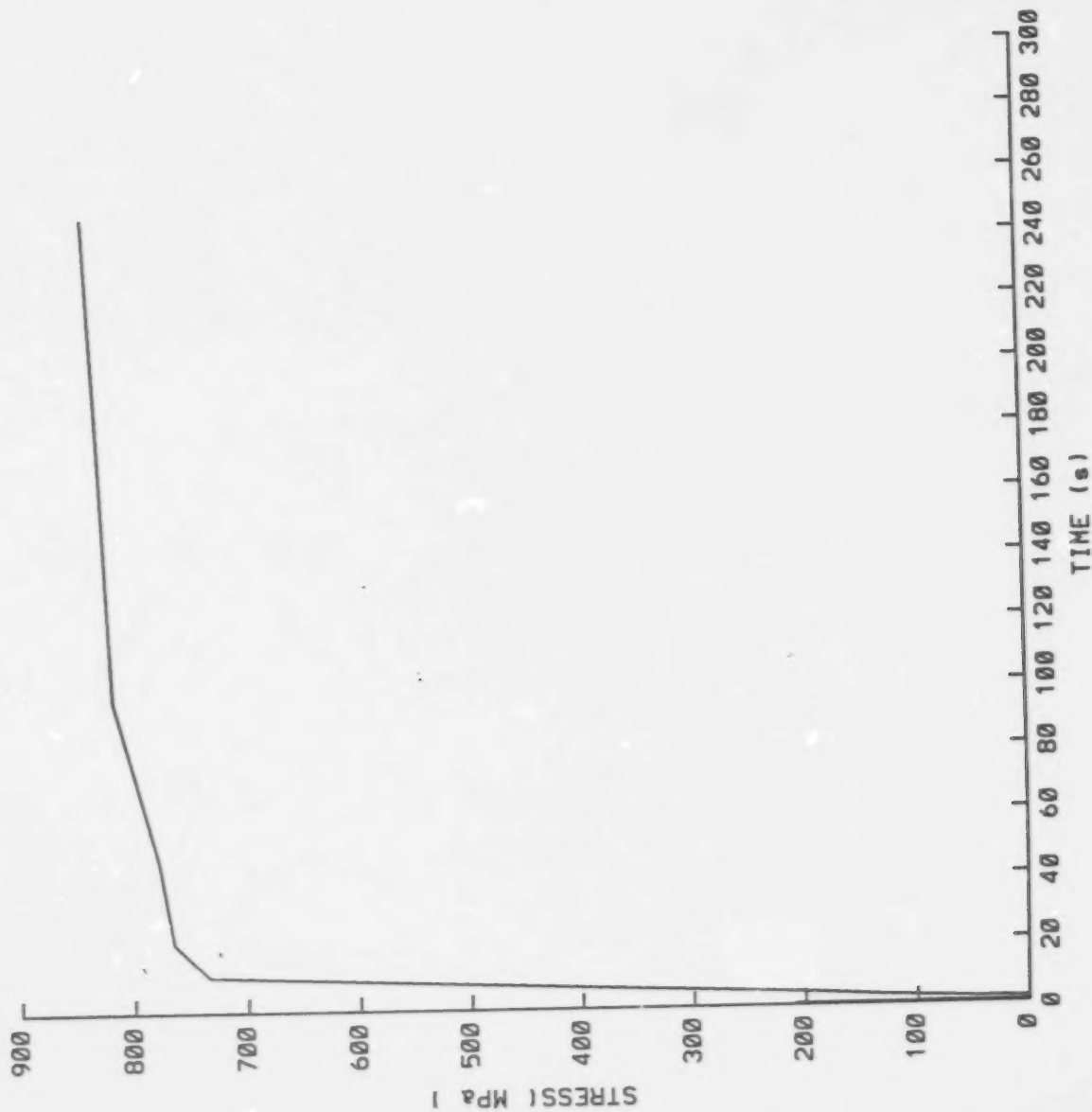


FIGURE 29: TRANSVERSE STRESS(FZ NØDE 68), RØØM TEMP. RL = 25 MM.

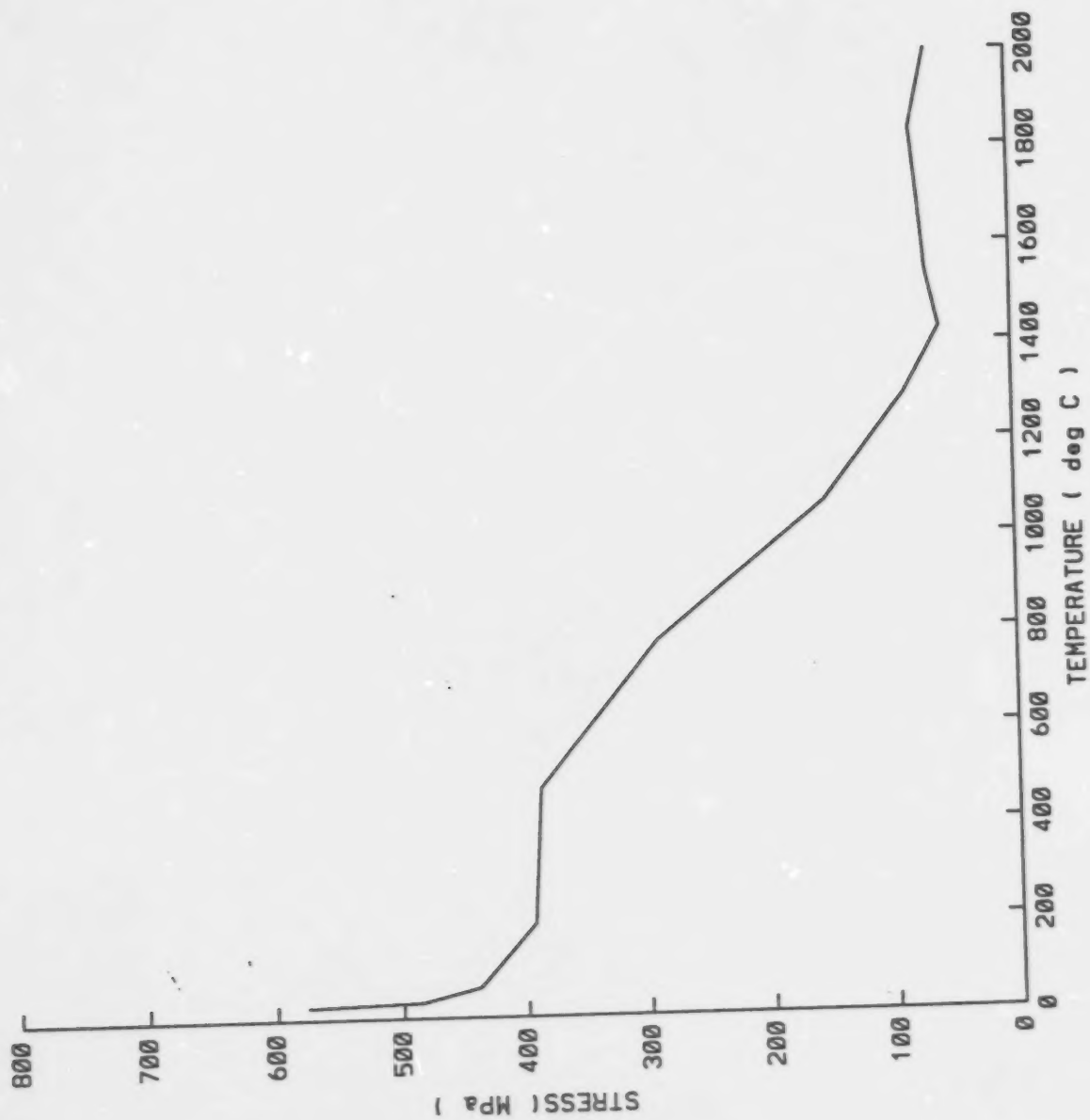


FIGURE 30: TRANSVERSE STRESS AT WELD CENTRE (TØP). RØØM TEMP. RL=100 MM.

SIGMAX - TEMP. PLOT

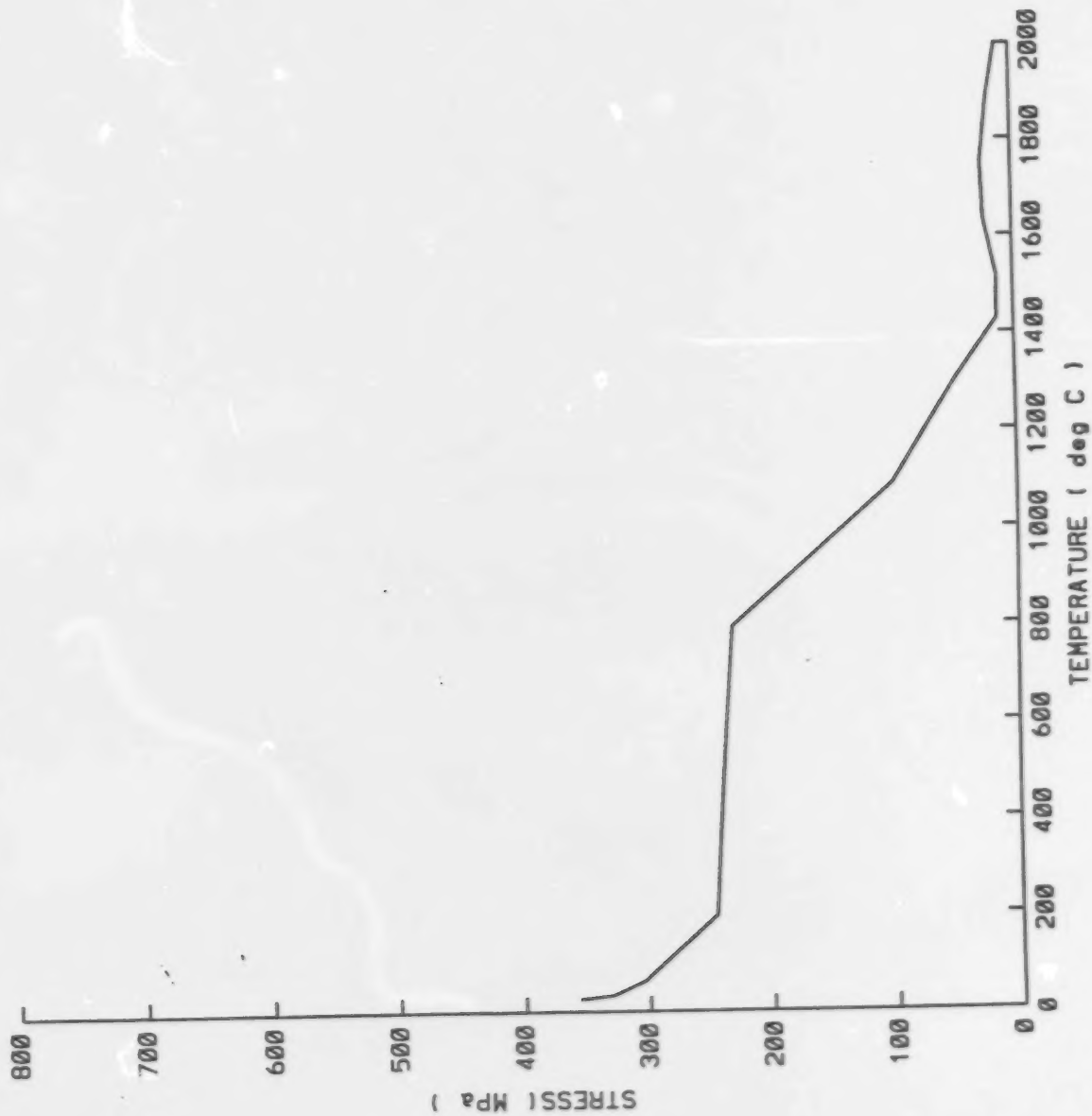


FIGURE 31: TRANSVERSE STRESS AT WELD CENTRE (BOTTOM). ROOM TEMP. RL=100 MM

SIGMAX - TEMP. PLOT

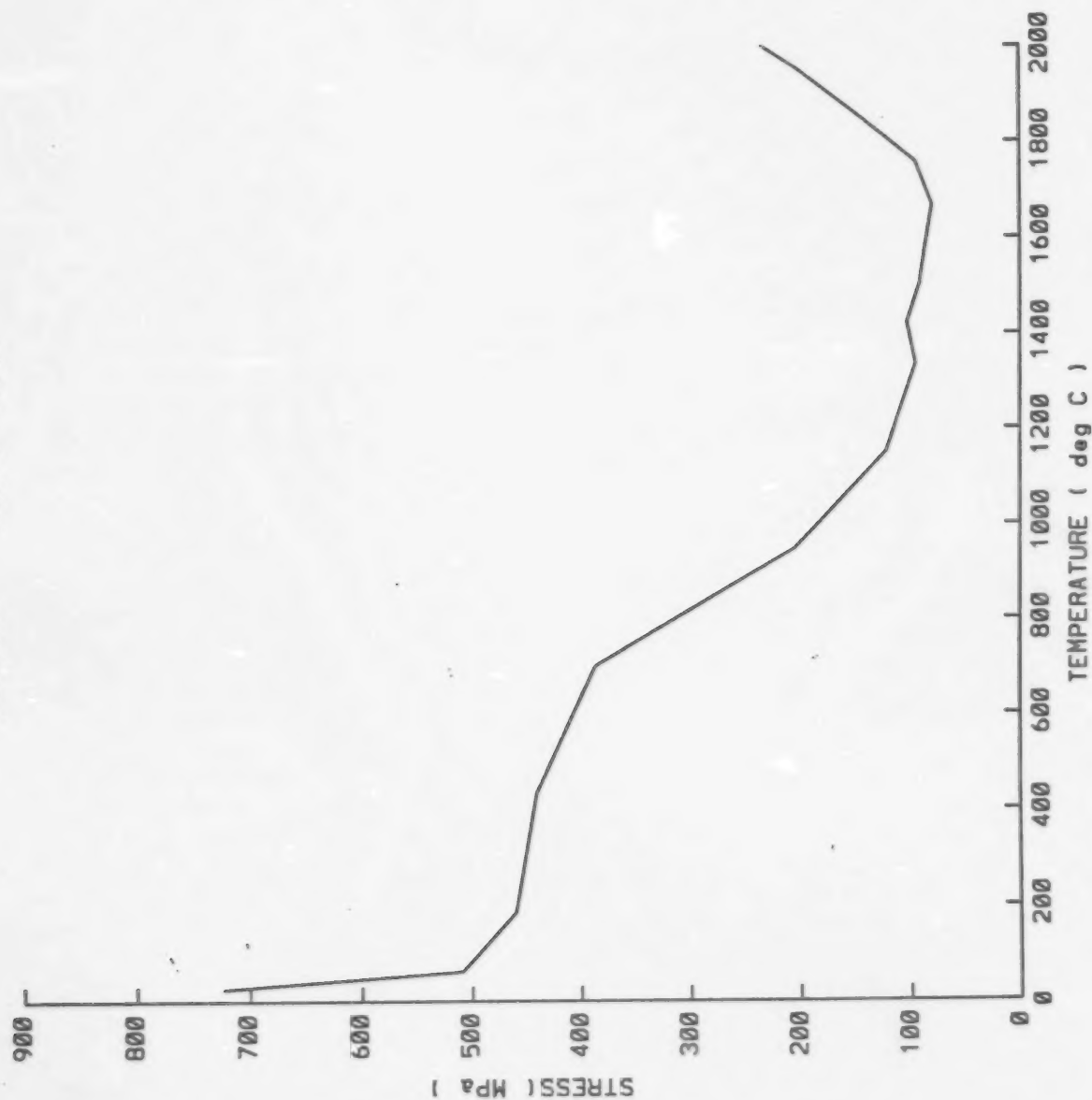


FIGURE 32: TRANSVERSE STRESS(NØDE 45). RØØM TEMP. RL = 100 MM.

SIGMAX - TIME PLOT

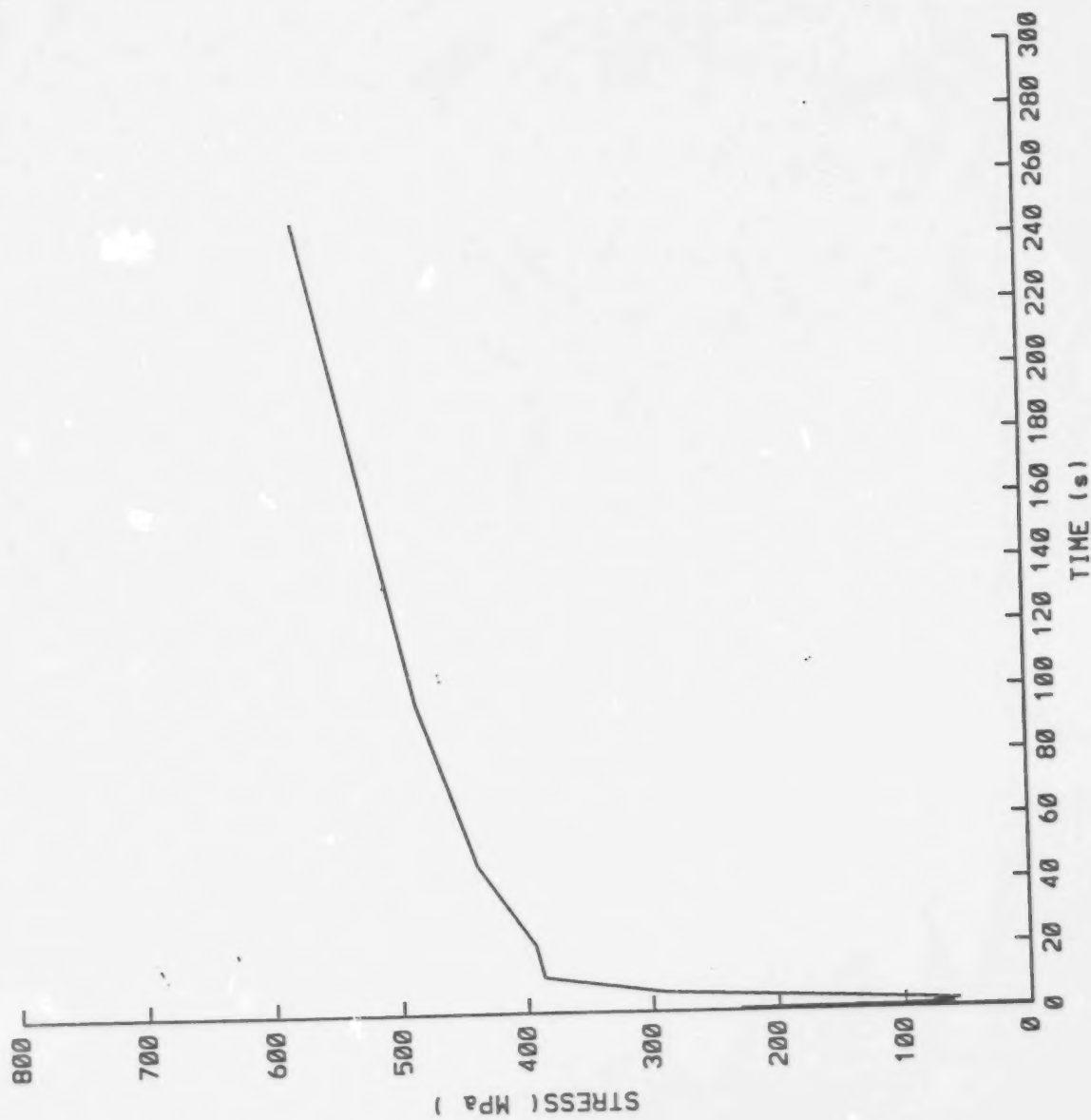


FIGURE 33: TRANSVERSE STRESS AT WELD CENTRE (TOP). ROOM TEMP. RL=100 MM.

SIGMAX - TIME PLOT

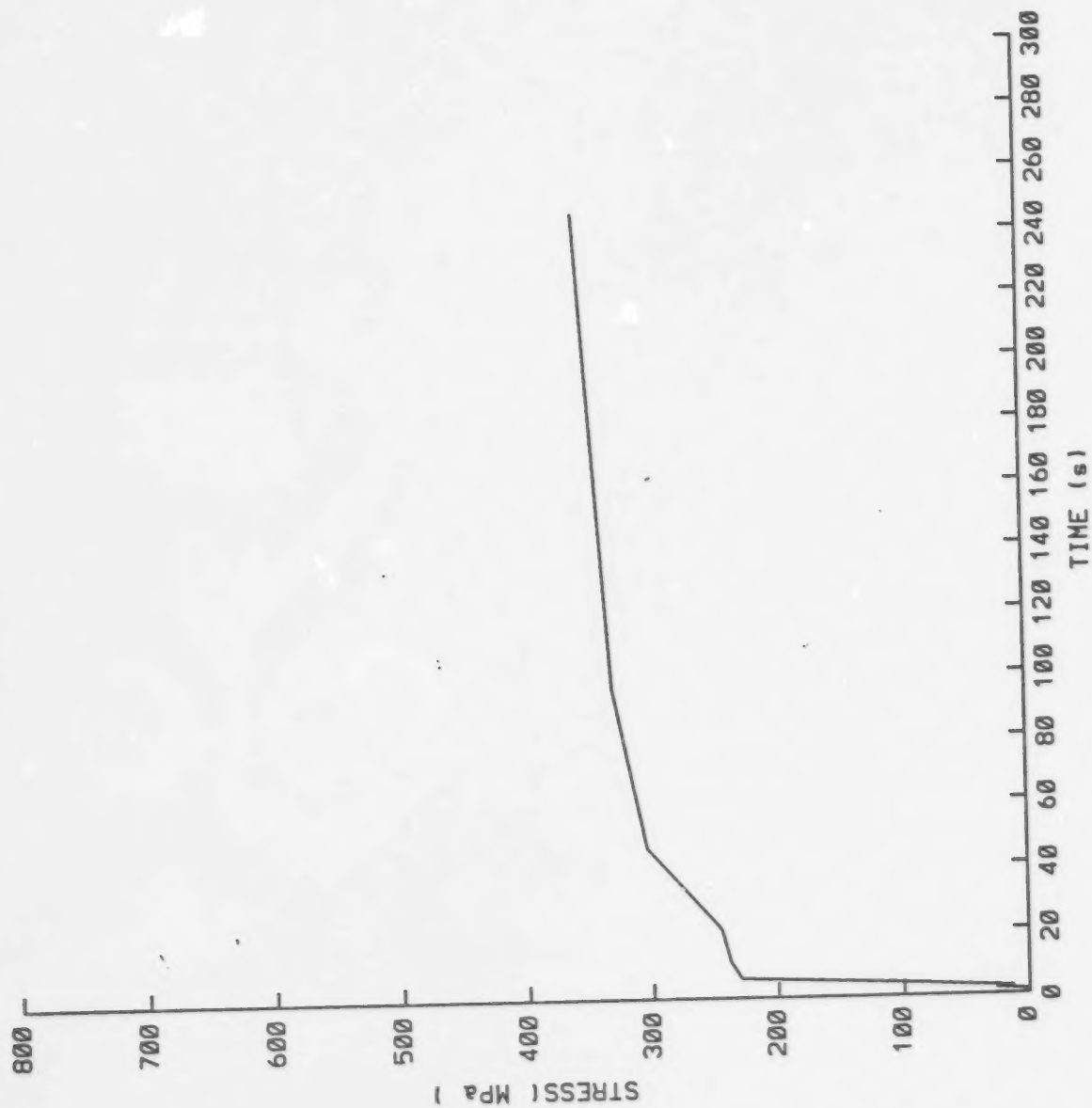


FIGURE 34: TRANSVERSE STRESS AT WELD CENTRE (BOTTOM), ROOM TEMP, RL=100 MM

SIGMAX - TIME PLOT

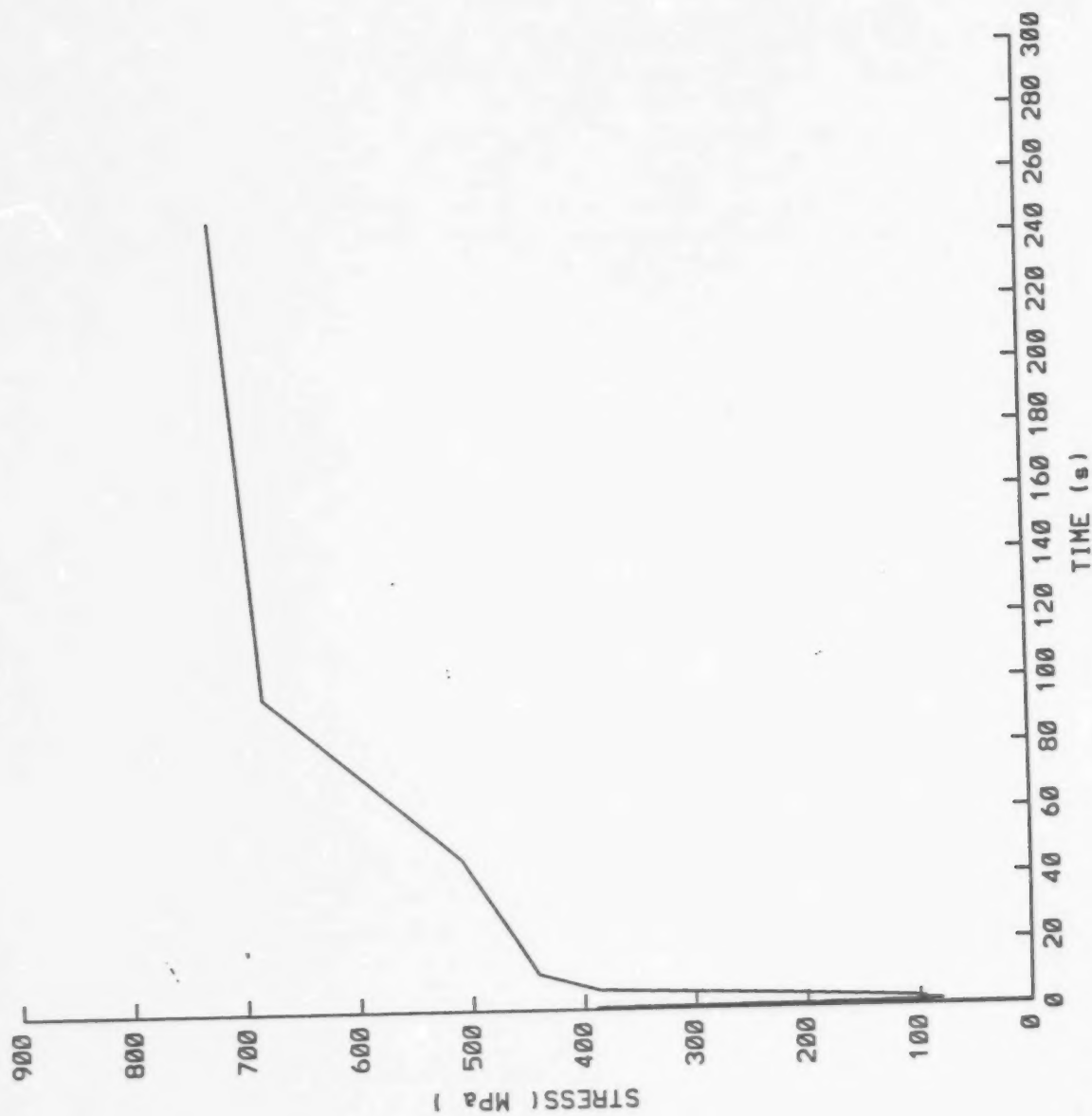


FIGURE 35: TRANSVERSE STRESS (NODE 45), ROOM TEMP. RL = 100 MM.

SIGMAX - TIME PLOT

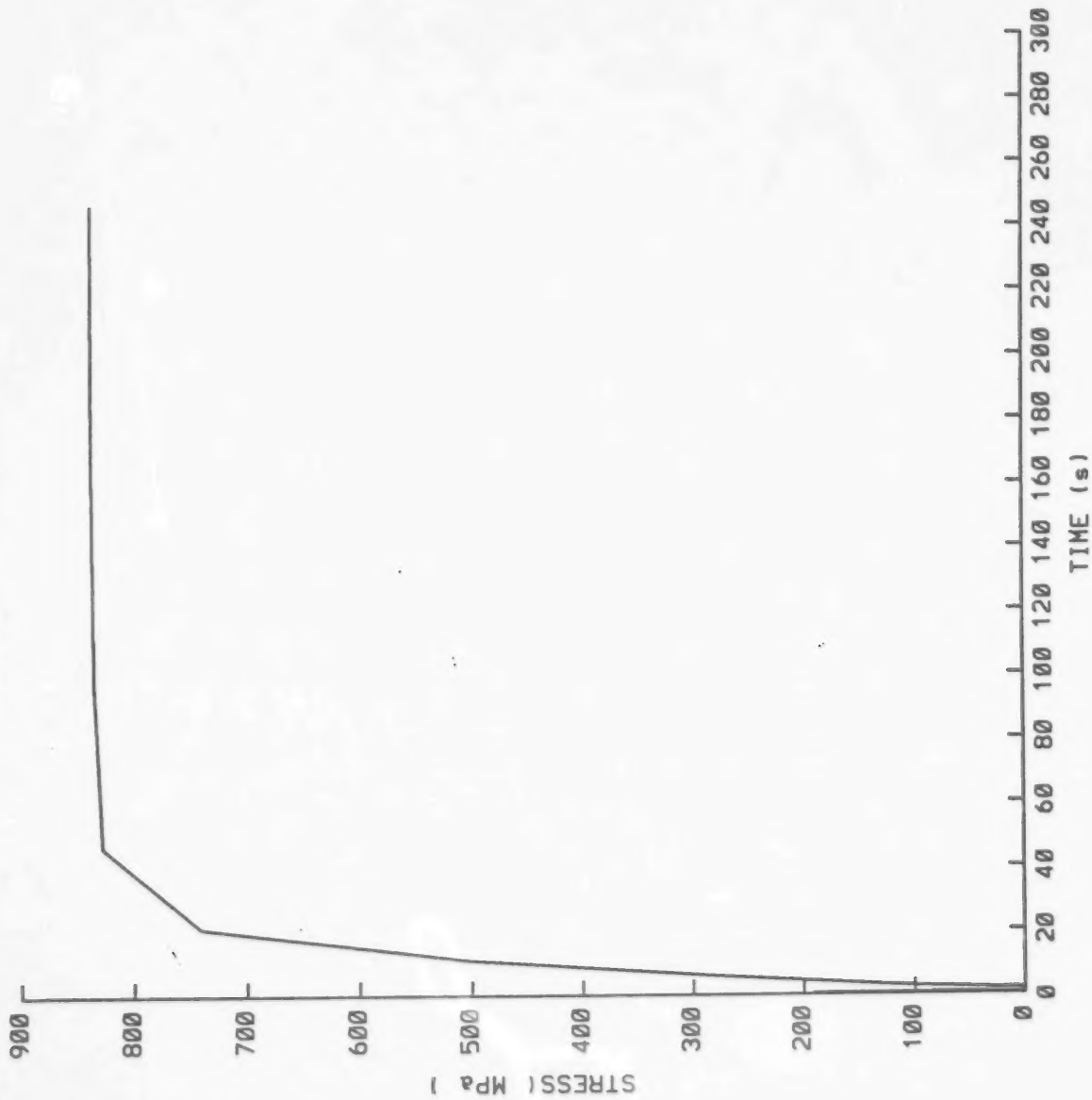


FIGURE 36: TRANSVERSE STRESS(FZ NØDE 68). RØØM TEMP. RL = 100 MM.

SIGMAX - TEMP. PLOT

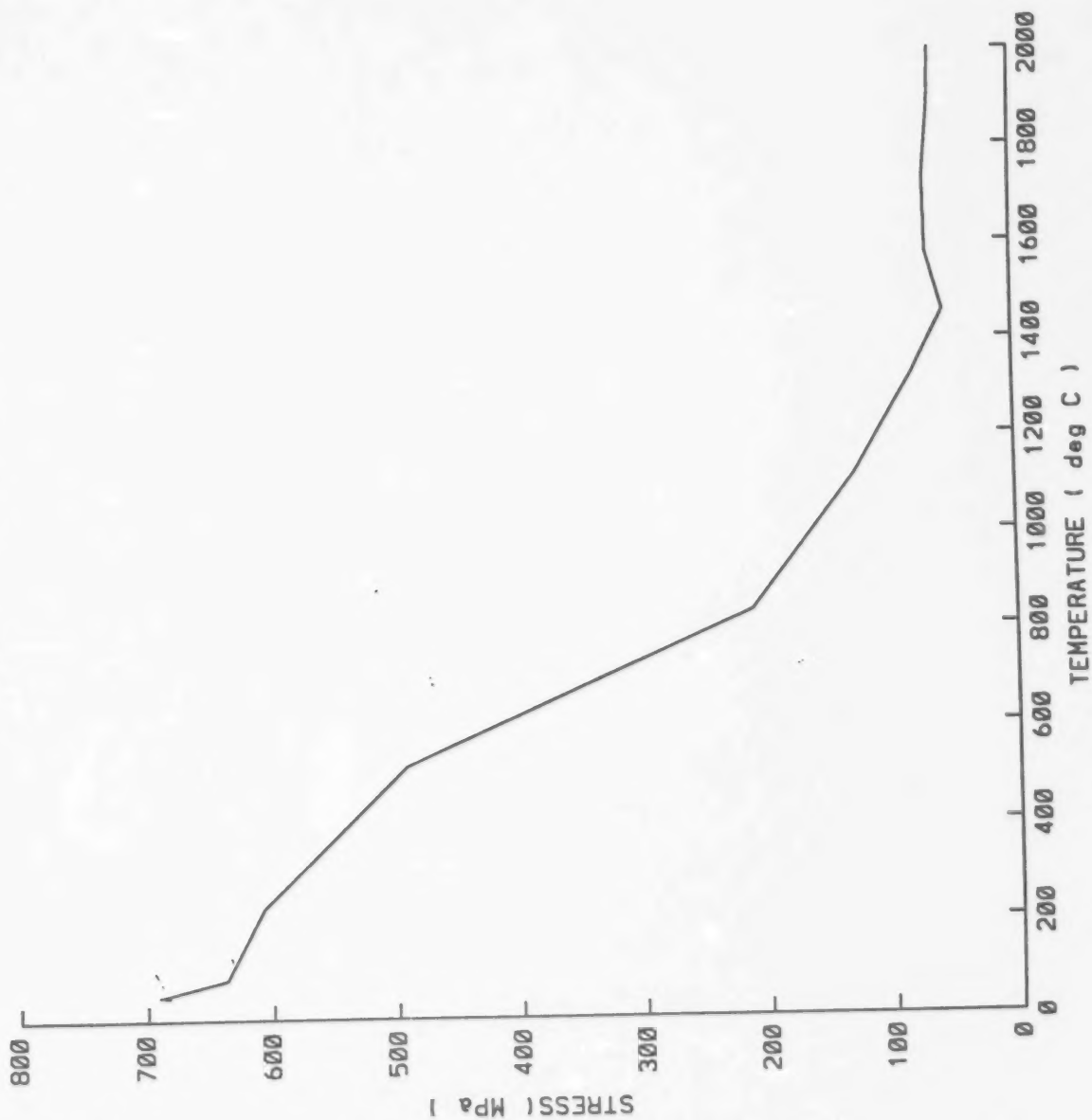


FIGURE 37: TRANSVERSE STRESS AT WELD CENTRE (TØP). PRHEAT=100. RL=25 MM.

SIGMAX - TEMP. PLOT

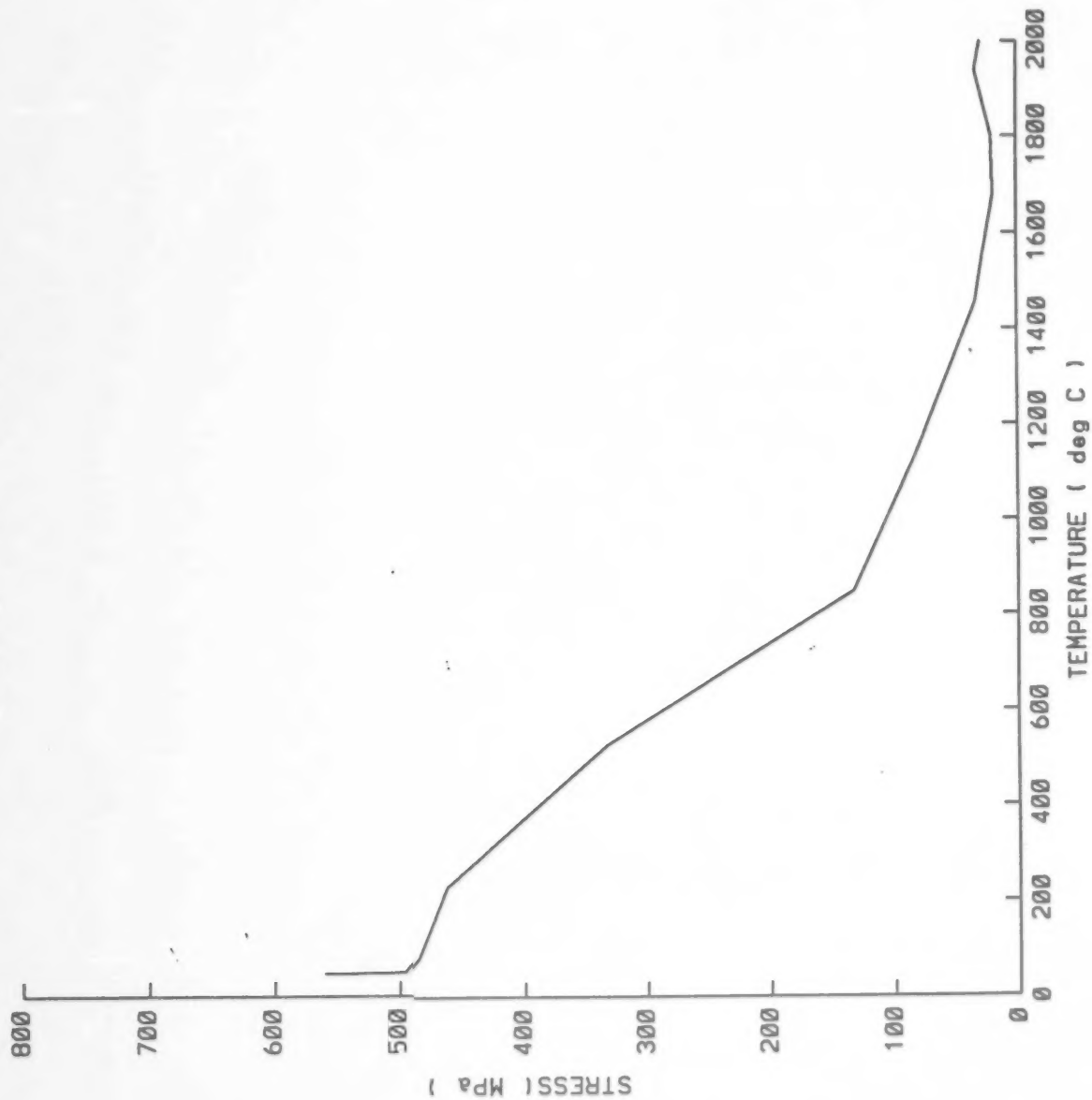


FIGURE 38: TRANSVERSE STRESS AT WELD CENTRE (BØTTØM). PRHEAT=100. RL=25 MM

SIGMAX - TEMP. PLOT

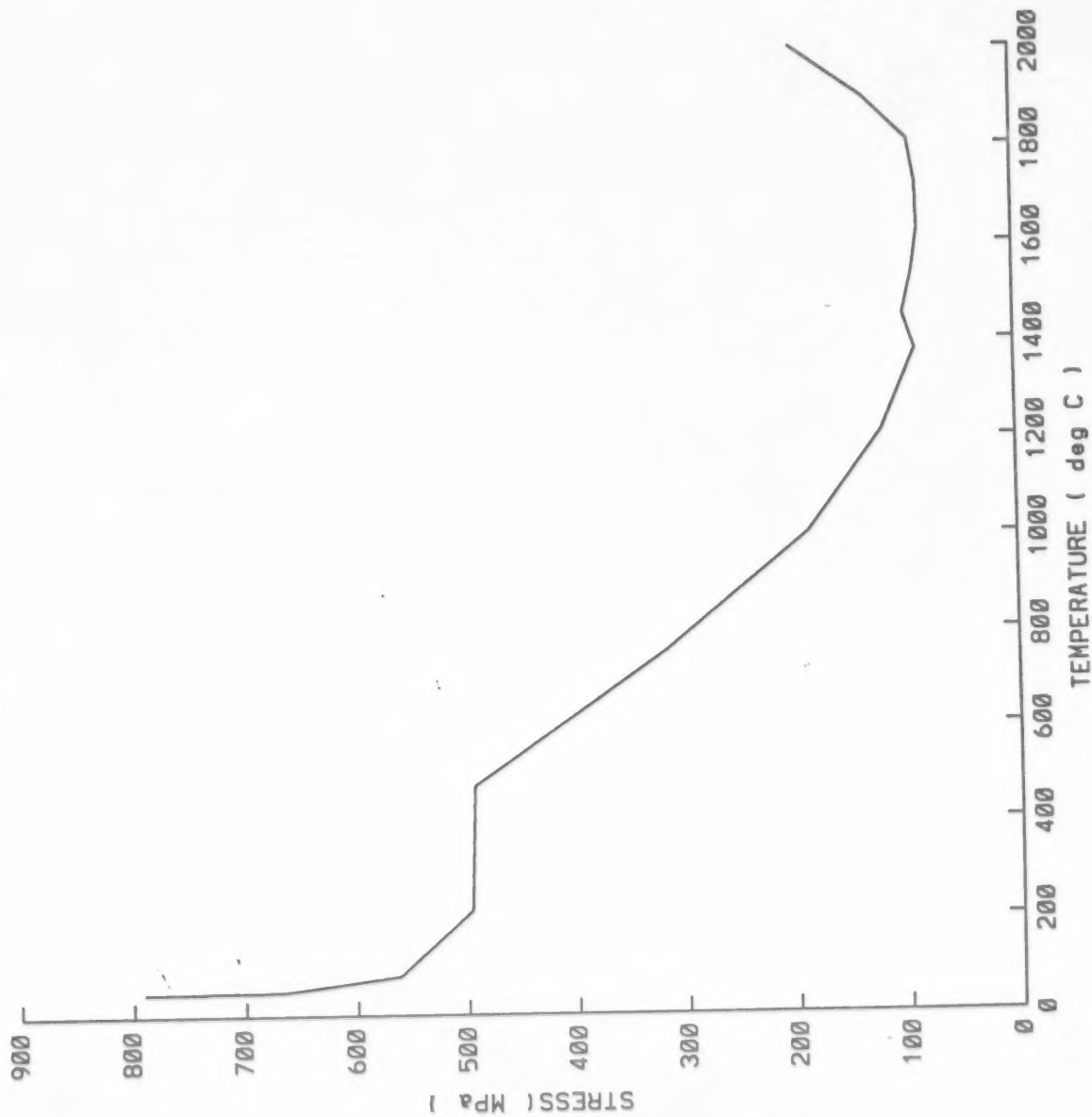


FIGURE 39: TRANSVERSE STRESS(NØDE 45).PRHEAT=100, RL = 25 MM.

SIGMAX - TEMP. PLOT

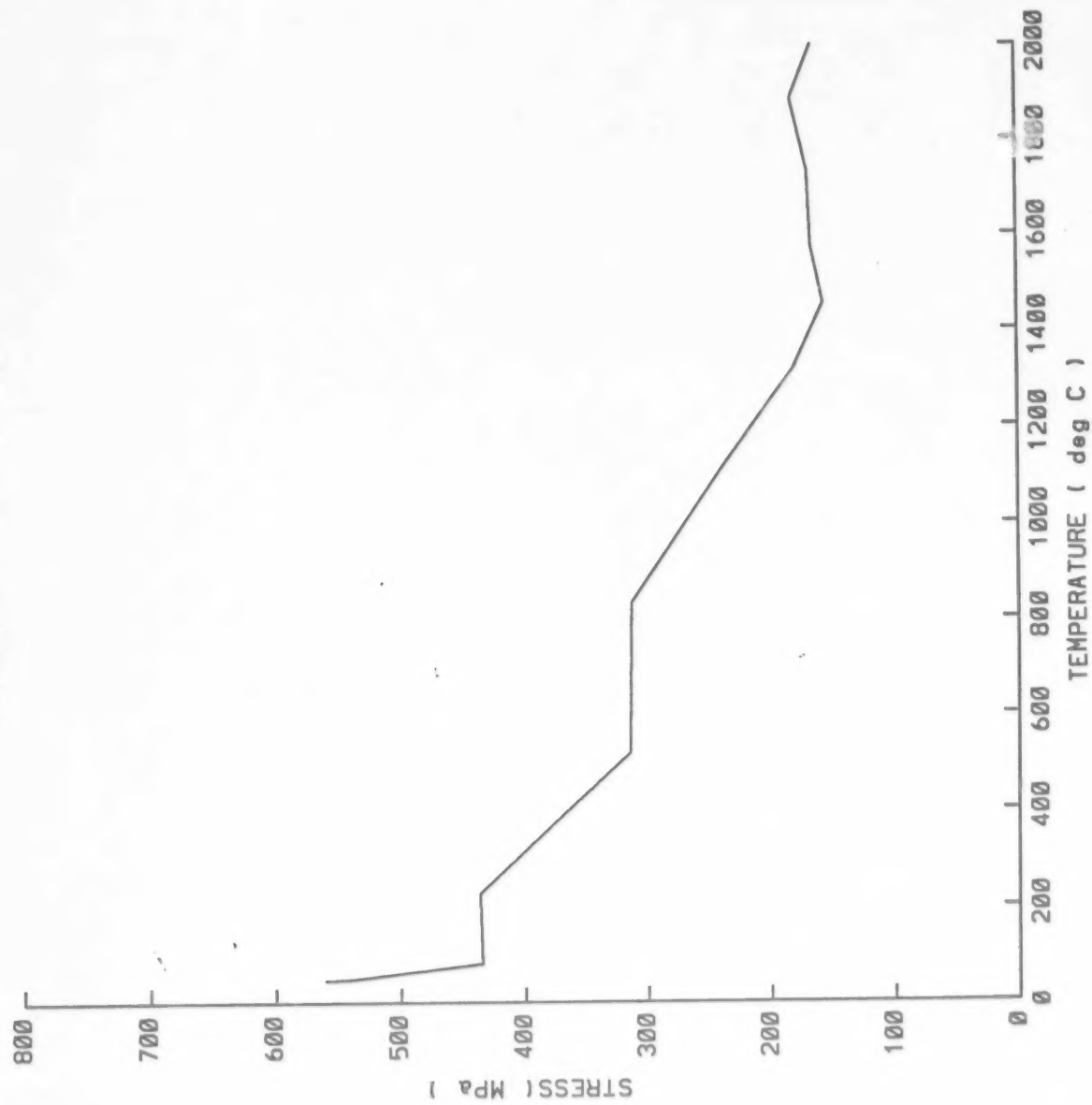


FIGURE 40: TRANSVERSE STRESS AT WELD CENTRE (TØP), PRHEAT=100, RL=100 MM.

SIGMAX - TEMP. PLOT

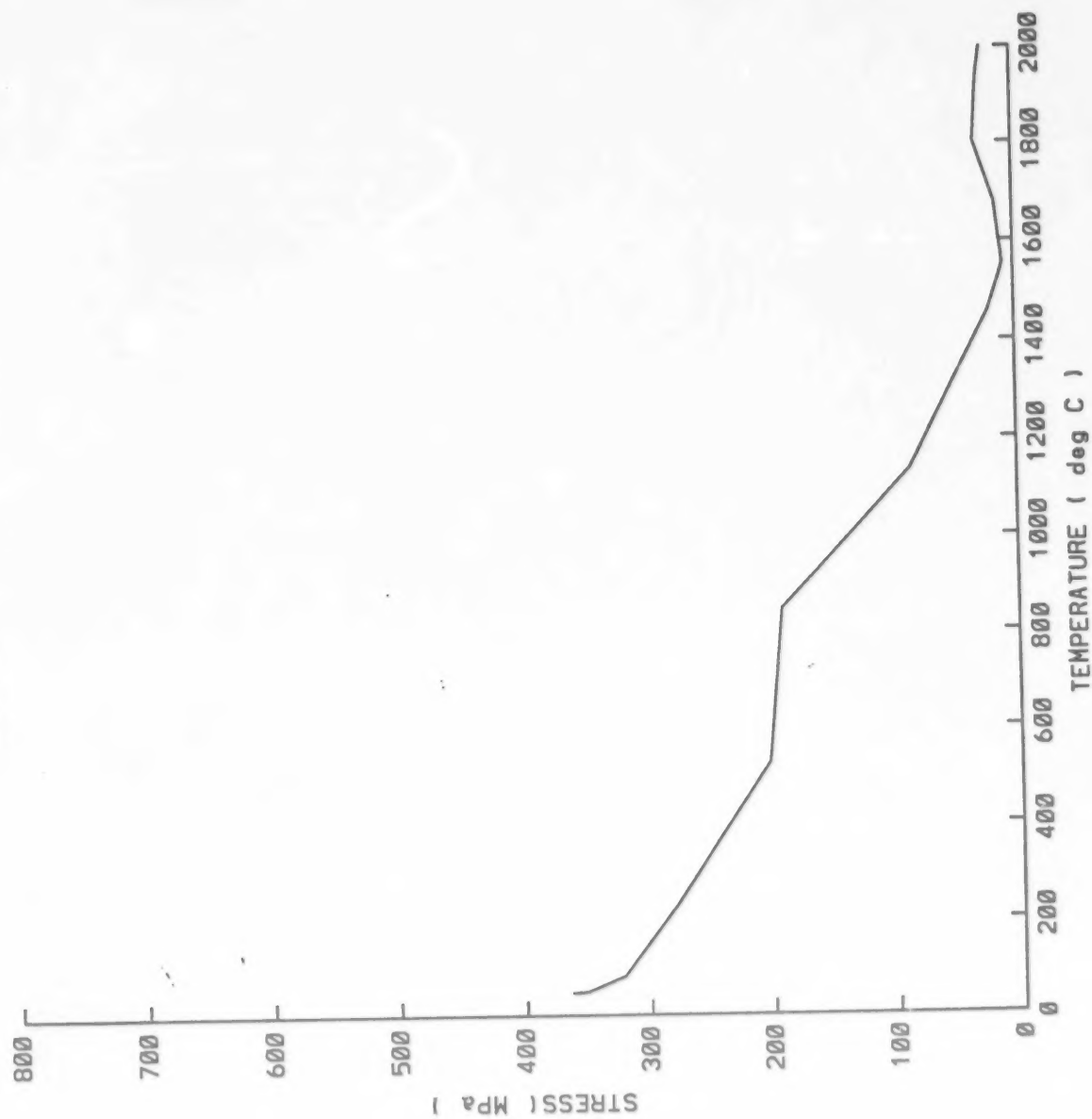


FIGURE 41: TRANSVERSE STRESS AT WELD CENTRE (BOTTOM). PRHEAT=100, RL=100 M

SIGMAX - TEMP. PLOT

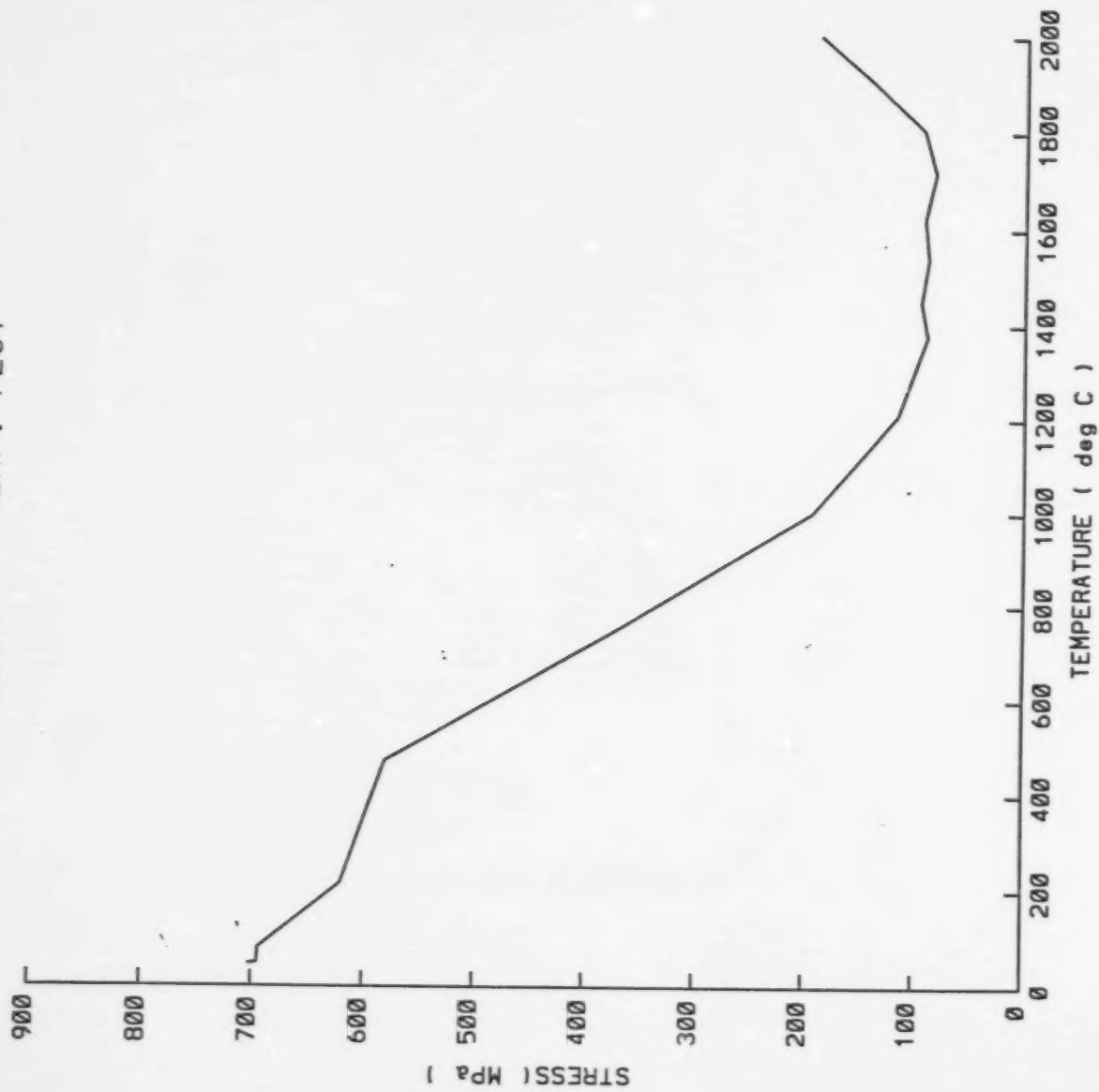


FIGURE 42: TRANSVERSE STRESS(NØDE 45).PRHEAT=100, RL = 100 MM.

SIGMAX - TIME PLOT

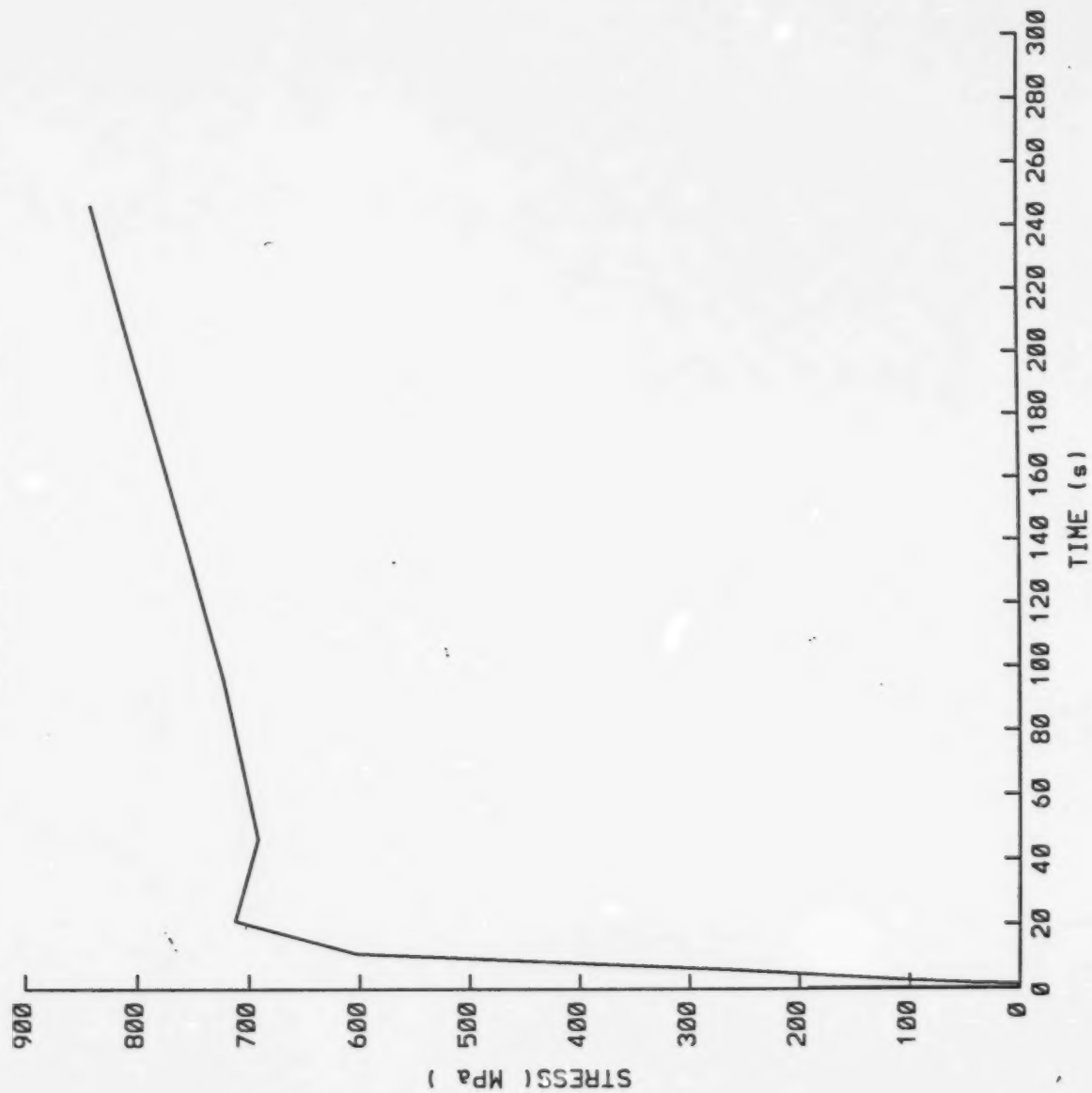


FIGURE 43: TRANSVERSE STRESS (FZ NØDE 68). PRHEAT=100. RL = 25 MM.

SIGMAX - TIME PLOT

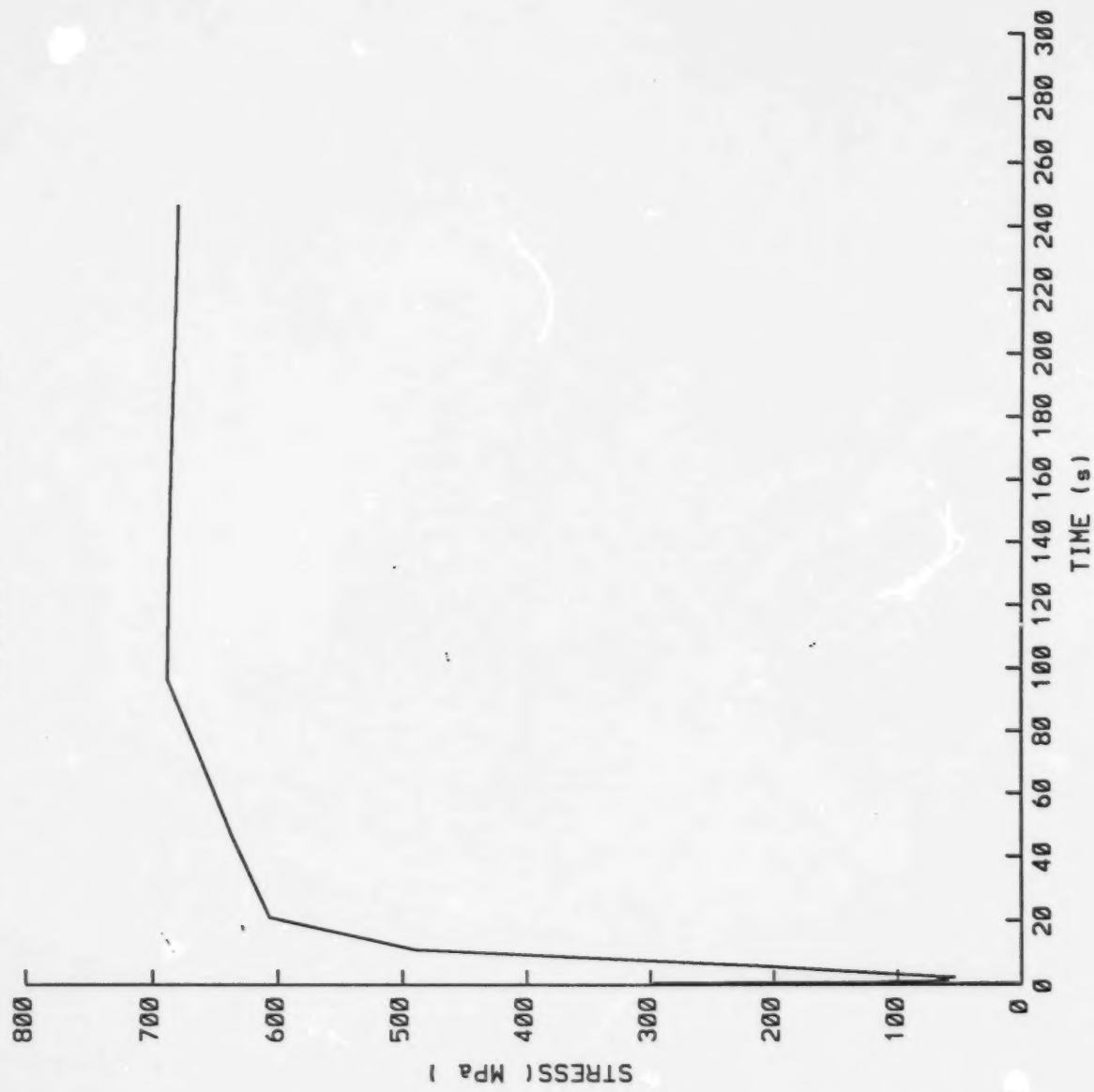


FIGURE 44: TRANSVERSE STRESS AT WELD CENTRE (TOP). PRHEAT=100, RL=25 MM.

SIGMAX - TIME PLOT

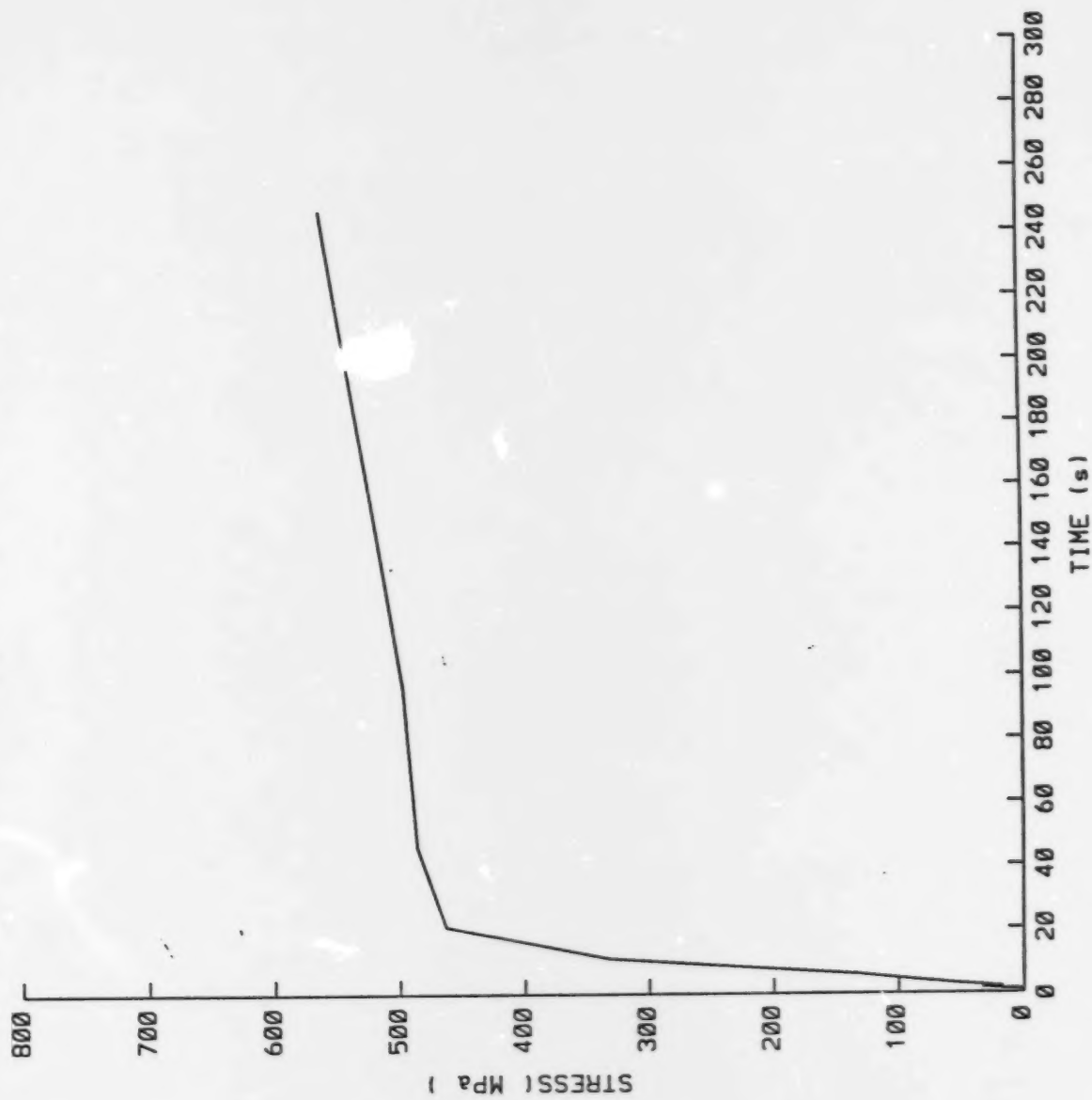


FIGURE 45: TRANSVERSE STRESS AT WELD CENTRE (BØTTØM). PRHEAT=100. RL=25 MM

SIGMAX - TIME PLOT

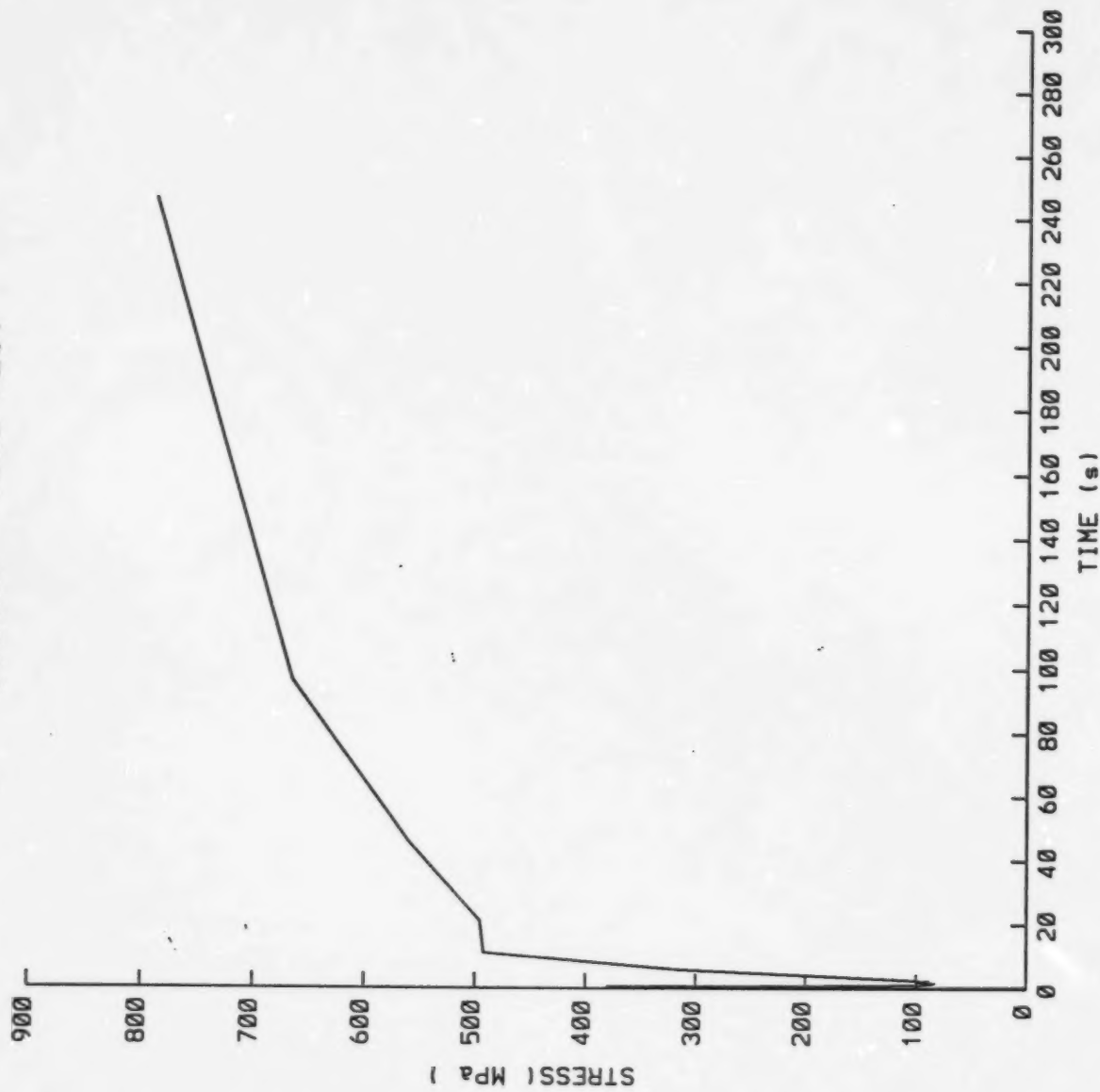


FIGURE 46: TRANSVERSE STRESS(NØDE 45).PRHEAT=100. RL = 25 MM.

SIGMAX - TIME PLOT

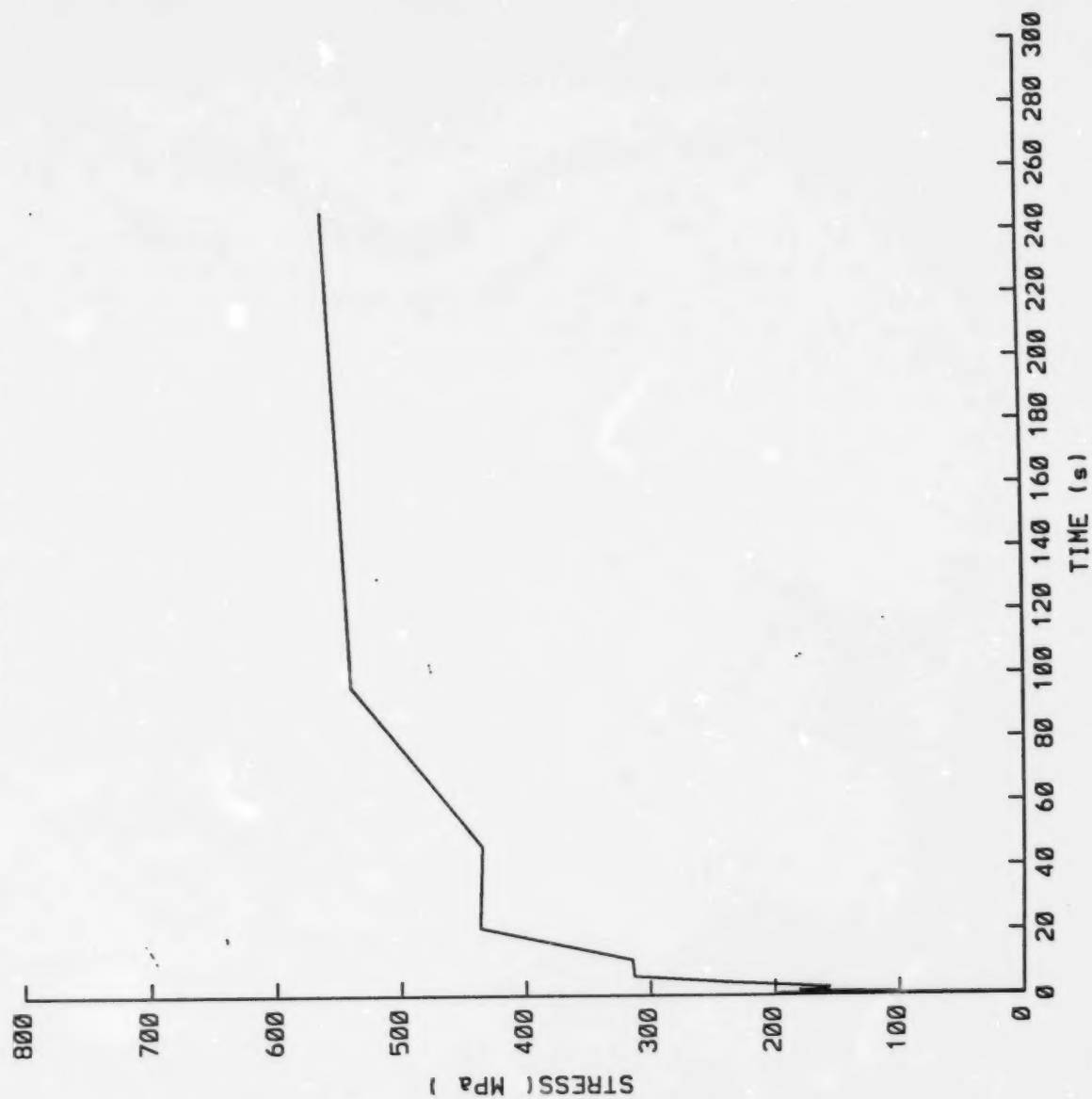


FIGURE 47: TRANSVERSE STRESS AT WELD CENTRE (TØP). PRHEAT=100, RL=100 MM.

SIGMAX - TIME PLOT

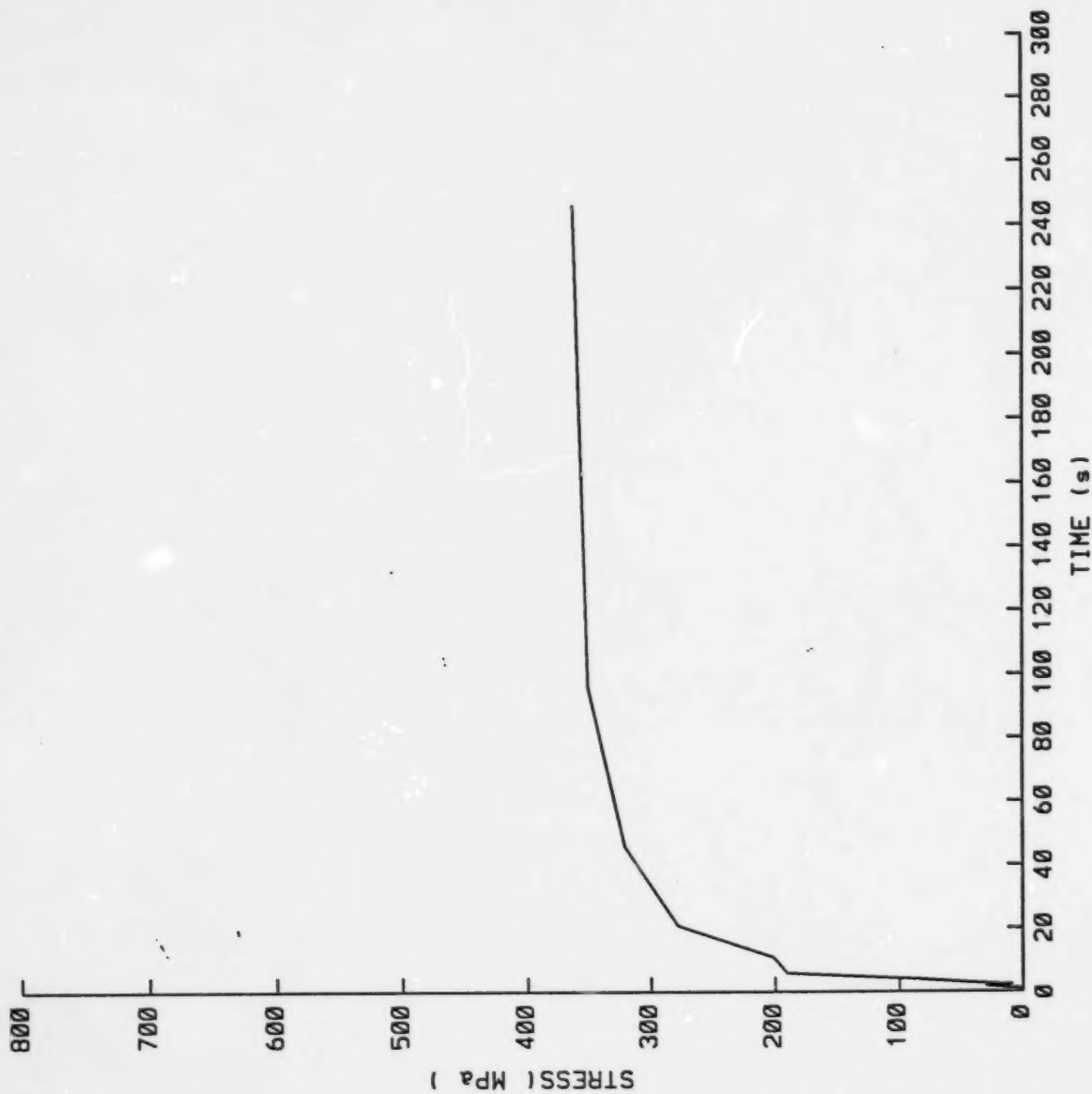


FIGURE 48: TRANSVERSE STRESS AT WELD CENTRE (BØTTØM). PRHEAT=100. RL=100 M

SIGMAX - TIME PLOT

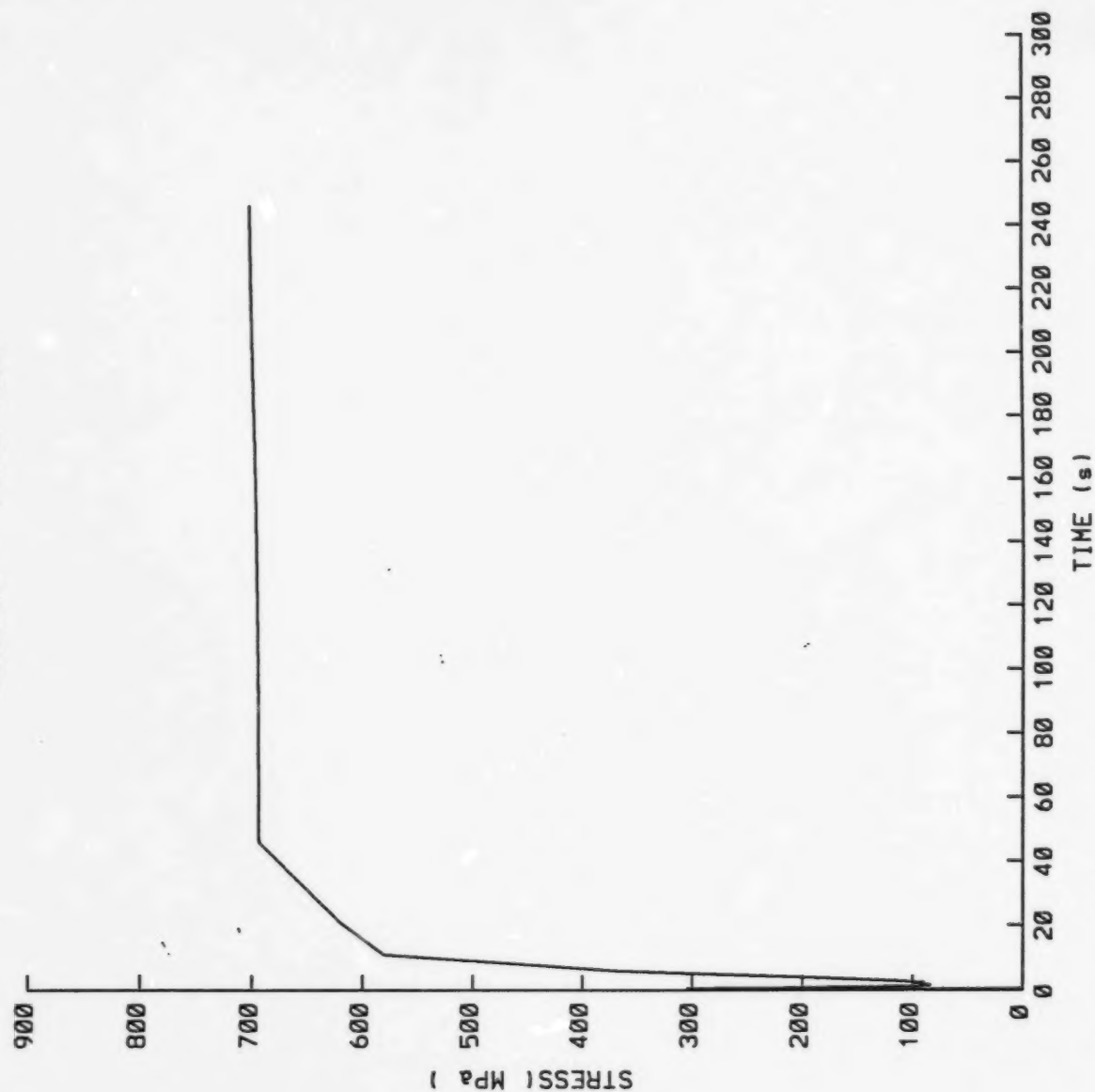


FIGURE 49: TRANSVERSE STRESS (NODE 45). PRHEAT=100. RL = 100 MM.

SIGMAX - TIME PLOT

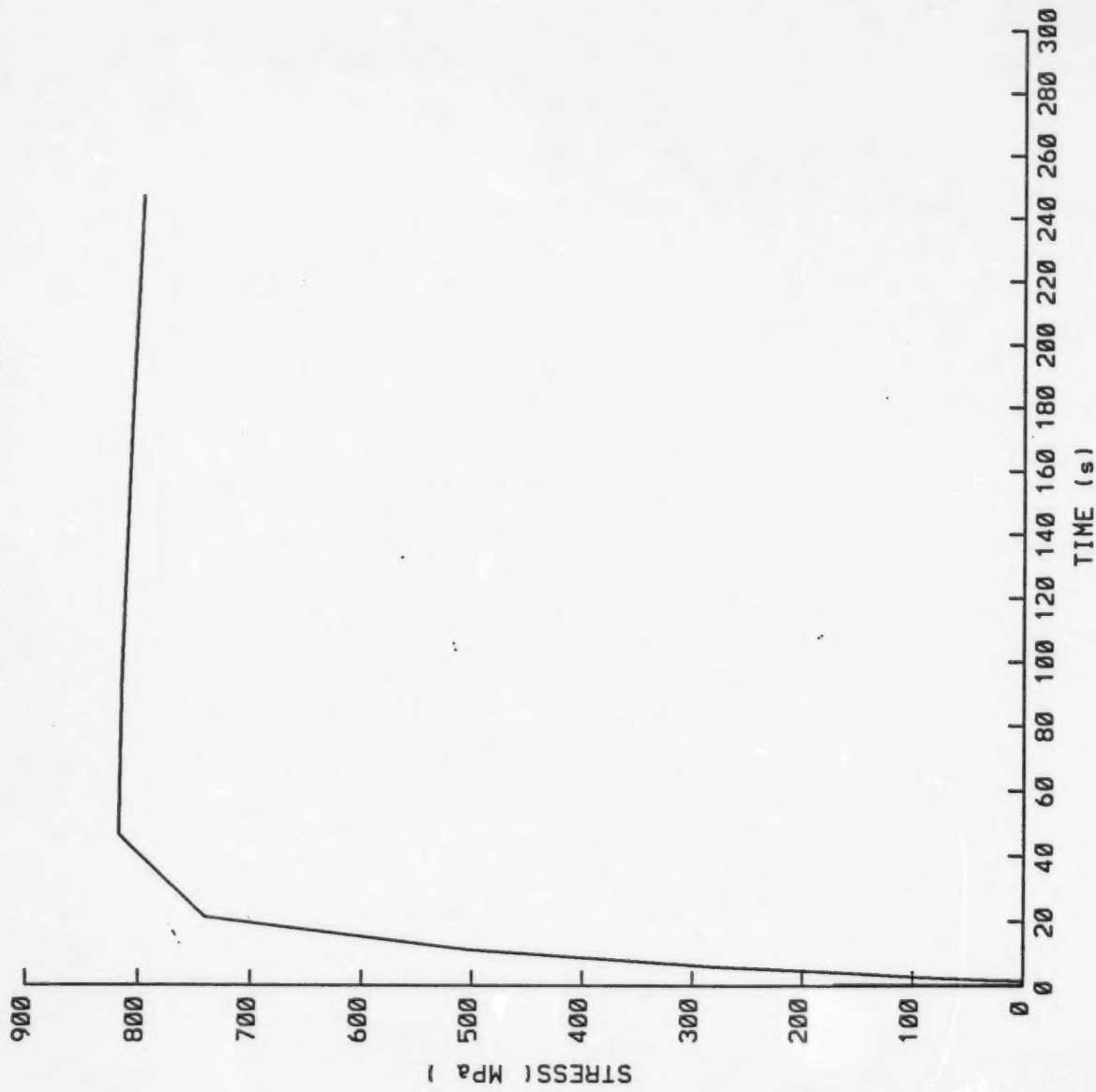
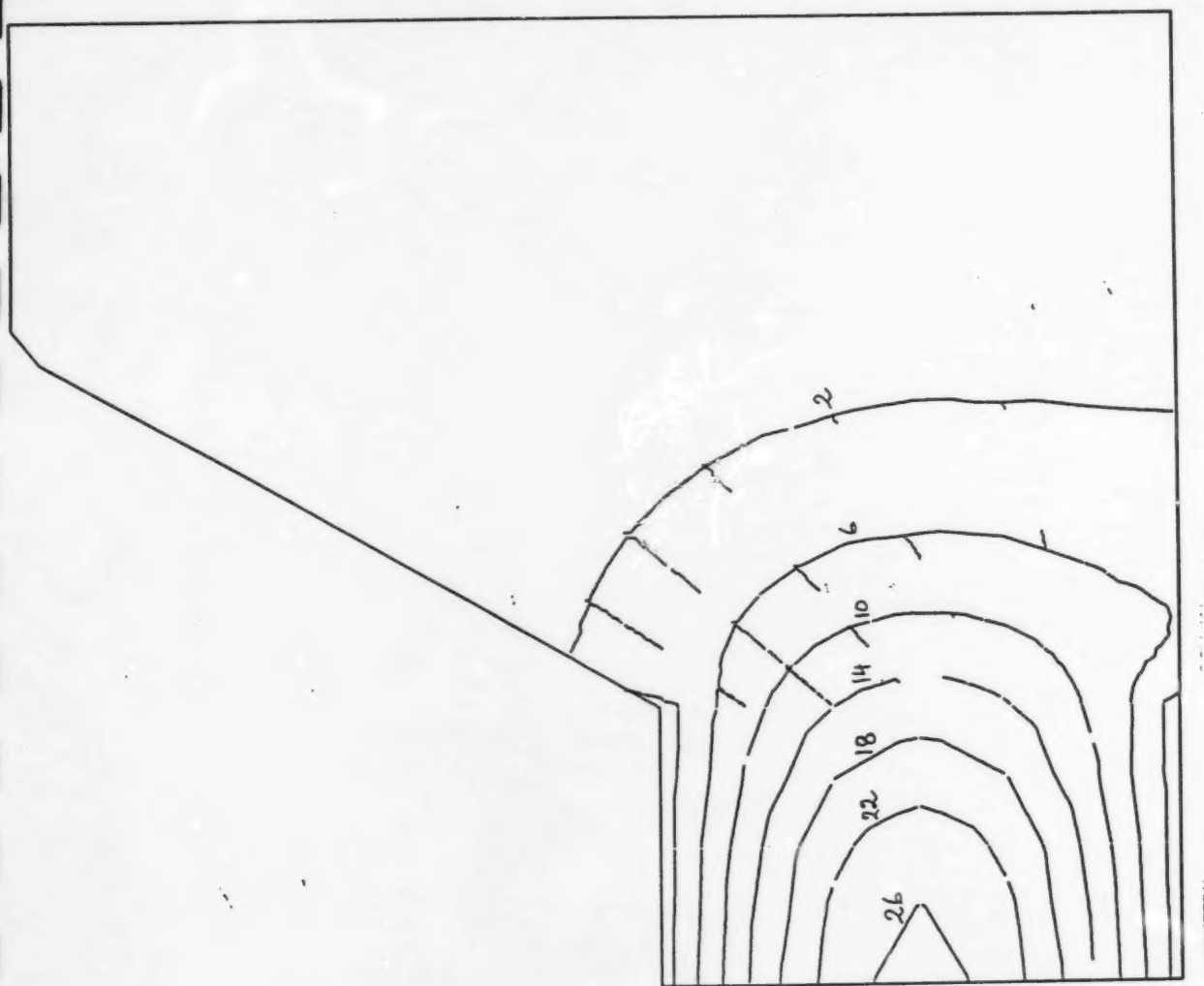
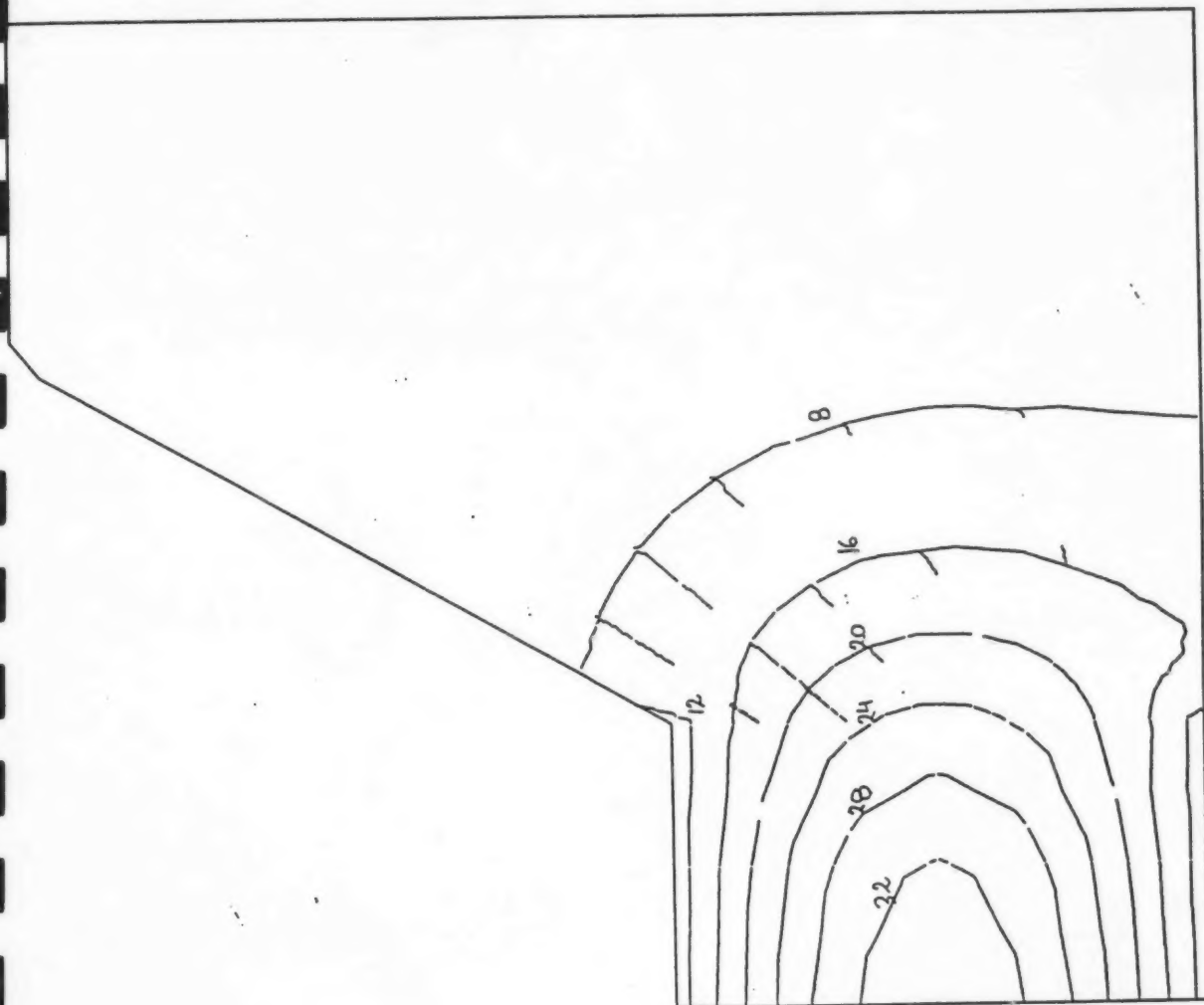


FIGURE 50: TRANSVERSE STRESS(FZ NØDE 68).PRHEAT=100. RL = 100 MM.

APPENDIX A



HYDROGEN CONTOURS. DIFFUSIVITY ZERO AT 700 DEG C.



HYDROGEN CONTOURS (DIFFUSIVITY ZERO AT 500 DEG C).